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REVIEW OF DIMENSIONAL INSTABILITY
IN METALS

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REVIEW OF DIMENSIONAL INSTABILITY IN METALS

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SUMMARY

This memorandum discusses some of the problems that arise as a result of dimensional instability, and presents data on stability, precision mechanical properties, and stabilization procedures for a variety of materials.

The term, dimensional instability, as it is used in this memorandum, refers to changes in dimensions that occur over a period of time in a specimen without external loading. The two primary mechanisms that cause dimensional instability in metals are (1) metallurgical instability and (2) relaxation of residual stresses. There are, in addition, more subtle metallurgical reactions that are not well understood. These may include the effects of ordering of interstitial and substitutional atoms, the effects of grain-boundary migration, and movements of magnetic domain walls. Some of the characteristics of the mechanisms leading to dimensional changes are discussed in this memorandum.

INTRODUCTION

Interest in the dimensional stability and in the precision mechanical properties of materials continues to run high. Therefore, this memorandum has been prepared. It is intended to supplement DMIC Memorandum 189 ("A Review of Dimensional Instability in Metals", by F. C. Holden, March 19, 1964). Information and ideas which have come to our attention since that time are included here. Much of the discussion is exactly as it was in the earlier report; however, none of the data included in the earlier report is repeated here.

BACKGROUND

The dimensional stability of a material refers to its ability to maintain its original size and shape over a period of time under specified environmental conditions. Although the term is self-explanatory, it becomes necessary not only to specify the conditions to which the material is exposed, but also the accuracy to which dimensional changes are measured. Because true dimensional stability can be defined as an absolute concept, it may be more realistic to consider the degree of instability that can be measured with suitable accuracy.

Improved techniques of metrology developed during the past decade or two have increased the potential accuracy of such measurements by one or two orders of magnitude. Similarly, the requirements of industry and government, as exemplified by the needs of missile and space systems, have become increasingly stringent. Manufacturing methods have been improved to the point where tolerances specified in microinches (millionths of an inch) are becoming commonplace; in many instances, it is important not only to manufacture a component with such precision, but also to ensure that its dimensions do not change during service. It may be expected that the standards for producing and maintaining very high degrees of precision in manufactured parts will continue to increase during the next decade, and that those will be extended into broader segments of industry not yet fully affected by the increased requirements for precision.

In the past, the distortion or dimensional instability of metals was studied mainly for the purpose of eliminating or reducing relatively large changes in dimensions in such parts as castings and die blocks. Most of these applications involved ferrous alloys, and a considerable volume of research was conducted to study the mechanisms leading to distortion, and methods for its reduction. A summary of the information available on this subject was presented in DMIC Report 163, "Control of Dimensions in High-Strength Heat-Treated Steel Parts".

Additional information of a somewhat different character is needed to meet material requirements for recent developments in precision devices, such as bearings, gyros, accelerometers, and missile-guidance systems. In these applications, very high degrees of precision and dimensional stability may be needed over long periods of time. The metals involved range from the more conventional alloy steels and aluminum alloys to the newer metals - titanium, beryllium, and the refractory metals. Interest also has been shown in composite structures (sandwich, laminates, etc.) and in nonmetallics - glass, ceramics, and plastics. In general, material selection is limited by factors other than dimensional stability; examples are strength/density, resistance to corrosion, elastic modulus, and magnetic behavior. The necessity for achieving specified physical or mechanical properties in addition to stability of dimensions frequently leads to difficulties, since the processing requirements often are incompatible.

Another problem area is involved with the conditions of service under which dimensional stability is to be maintained. The influence of temperature and stress, both steady and cyclic, combined with the presence of various types of fields are the most important variables. A part of the dimensional change is (in most materials) unavoidable but predictable: thermal expansion and contraction from temperature changes, and elastic strain from stress application, for example. These effects usually can be compensated for by suitable design, and can be minimized by careful selection of material. For example, the thermal expansion can be reduced to essentially zero over a restricted temperature range by selecting a suitable alloy of the Invar type. Elastic strains can be minimized by using a material with a high elastic modulus, and by designing for low stress levels. The thermal-expansion and elastic-strain effects are essentially reversible, and are not ordinarily considered as a form of dimensional instability.

Many of the available data have been obtained on specimens that are not subjected to external loads other than their own weight. This probably is because much of the initial research in this field was done to develop improved methods for making reference standards, such as gage blocks,

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rather than components subject to external loads. On the other hand, most parts in precision equipment are subjected to stress during service, even though the stress levels usually are relatively low. It has been observed that deformation, both time independent and time dependent, can occur at the microinch-per-inch level at stresses well below the conventional yield stress or proportional limit. As an example, the conventional yield strength (0.2 percent offset) for wrought 6061 aluminum alloy was reported to be 40,000 psi, whereas the precision elastic limit was about 12,000 psi.

Studies on the mechanisms of microstrain have been carried out rather intensively in recent years. Although terminologies vary, the terms "precision elastic limit" and "microcreep limit" have been used to designate the stresses at which time-independent and time-dependent plastic flow occur. The precision elastic limit (PEL) is defined as the lowest stress at which a specified residual strain (usually of 1 microinch per inch) is detected. It is ordinarily determined by loading to successively increasing stresses in tension until a residual strain is detected (Figure 1)*. The microcreep limit, as defined by Hughel,(1)** is the lowest stress sufficient to cause a progressive increase in residual strain on three successive loadings.

As a result of the foregoing, it appears that in a stressed part, the importance of microstrain, as distinguished from true dimensional instability, must be recognized. For convenience, therefore, the total dimensional change is considered to be composed of three parts:

(1) Recoverable dimensional changes; time independent (these generally are understood and predictable, and include elastic strain, thermal expansion, and magnetostrictive strain) and time dependent (these include stress-induced and magnetically induced ordering).

(2) Plastic deformation (microstrain); this term includes the irrecoverable plastic strains, time dependent and time independent, that result from applied stress.

(3) Dimensional instability; this term is reserved here for changes in dimensions resulting from internal stress systems or metallurgical instability (such as precipitation or phase changes); that is, changes that occur in the absence of external forces.

In the discussions that follow, these three causes of dimensional change are discussed, and available information on how they can be controlled is presented. Emphasis is placed upon the causes and effects of dimensional instability.

RECOVERABLE DIMENSIONAL CHANGES

Certain generally considered recoverable dimensional changes result from external changes in stress, temperature, and magnetic fields. Both linear and volume changes are involved. The elastic modulus (E) relates the magnitude of the applied stress to the corresponding elastic strain; the coefficient of linear expansion (α) relates the change of temperature to the resulting thermal strain, and the joule magnetostriction coefficient (λ) relates the magnitude of an applied magnetic field to the corresponding linear dimensional change.

Within restricted ranges of temperature and tolerance, these dimensional changes can be considered calculable and reversible, for these parameters (E, α, λ) are usually expressed as constants. It is well known, however, and must be remembered, that they are really average values. Moreover, there is usually a difference in the strain path, depending upon whether the applied force, temperature, or field is increasing or decreasing. This path difference traces out a hysteresis loop as, for example, shown in Figure 2.

A second example, for zirconium is given in Figure 3. These data emphasize the magnitude attainable by the hysteresis loop and the significance this behavior may often have in precision mechanical-property considerations. Time-dependent after effects are often associated with such hysteresis behavior, as shown in Figure 4. Data such as these lead to the generalization that plastic deformation, in the absence of stress relief, often (perhaps always) adversely affects the dimensional stability.

The importance of using caution when defining various precision mechanical properties, and when interpreting data from other sources cannot be overemphasized. For example, some authors prefer to define elastic limit as the lowest stress at which the hysteresis loop is observed on a cyclic stress-strain curve. But <u>all</u> known stress-strain curves exhibit hysteresis. This is the basis of the whole field of internal friction. Therefore, such data must be considered in relationship to the sensitivity of the experimental equipment involved.

These so-called reversible effects can be predicted and, to a degree, minimized individually, provided that other considerations do not preclude a free choice of material and condition. The three effects described here frequently are, to a degree, related. For example, alloys of the iron-nickel type designed for low coefficient of expansion depend upon magnetostrictive effects to accomplish this, as do similar alloys with a controlled variation of elastic modulus with temperature. Invar and Ni-Span-C are two examples of such alloys.

For most purposes, the conventional handbook values for the parameters E, α, λ, are sufficiently accurate to provide design information. Where greater accuracy is needed for a specific application, it probably will be necessary to conduct experiments on the particular material and condition to be used, since variations in composition and structure are likely to be significant.

PLASTIC DEFORMATION (MICROSTRAIN)

As it is employed here, the term microstrain is defined as the irrecoverable plastic strain resulting from an applied stress. It has been pointed out for many years that the values of the elastic limit and the proportional limit of a metal, as conventionally defined, depend upon the precision of the strain measurement. Advances in measurement techniques now have progressed to the point where residual strains can be measured to the

* Figures begin on page 9.
** References are given on pages 4 and 5.

resolution of better than 1 x 10^{-6} with resistance strain gages, and to 1 x 10^{-7} or 1 x 10^{-8} by suitable capacitance gages. Etch pit and internal-friction techniques permit even greater solution (see Figure 5).

The precision elastic limit (PEL) is the stress at which the residual strain is 1 microinch per inch. This rather arbitary number is a simple recognition of the fact that, as shown in Figure 1, the ultimate strength, or even the yield point is far too coarse a number for the designer of precision equipment to use. A comparison of some of these numbers for various materials (see Tables 1-5* and Figures 6-15) emphasize this point.

The residual microstrains corresponding to the precision elastic limit or the anelastic limit are considered to be essentially time independent. Studies of time-dependent deformation at microstrain levels also have been conducted. The term "microcreep limit" has been defined by Hughel[1] as the stress just sufficient to cause progressive increase in residual strain on three successive loadings to the same stress level. For beryllium, it was found that the microcreep limit was significantly higher than the precision elastic limit. In other work, microcreep in Invar and 356-T6 aluminum at room and slightly elevated temperatures has been observed at stresses near (and in some instances below) the elastic limit. Various examples are given in Figures 16-31.

DIMENSIONAL INSTABILITY

The term "dimensional instability", as it is used here, refers to changes in dimensions that occur over a period of time in a specimen without external loading. Data have been reported for a number of metals and alloys exposed both at constant temperature and to temperature cycling.

Mechanisms Leading to Dimensional Instability in Metals

The two primary mechanisms that cause dimensional instability in metals are reasonably well known. These are (1) metallurgical instability and (2) relaxation of residual stresses. There are, in addition, more subtle metallurgical reactions that are not so well understood. These may include the effects of ordering of interstitial and substitutional atoms, the effects of grain-boundary migration, and movements of magnetic domain walls. The effects of radiation on dimensional changes and on properties of materials, particularly fuel element materials, have been studied extensively; however, these are considered to be beyond the scope of this memorandum. Some of the characteristics of the mechanisms leading to dimentional changes are discussed in the following sections.

Metallurgical Mechanisms

(1) Metals or alloys that do not undergo a phase change form one of the simplest classes of materials. The only apparent microstructural changes are in grain size, shape, and orientation. One metallurgical change which can cause small dimensional changes is ordering. Individual solute atoms often will tend to occupy specific positions in the solvent lattice relative to like or unlike atoms. Because these reactions are controlled by the diffusivity of the solute in question, the reaction rates are distinguished by a relatively strong temperature dependence. Small dimensional changes will follow changes in stress, magnetization, or possibly temperature. Such reactions can be responsible for warm-up times for oscillating devices, hysteresis behavior during the stress cycle, or time dependence **after** reaching some fixed new temperature.

(2) An alloy that rejects a second phase from solid solution (typical of the age-hardening alloy systems) will usually undergo a gradual change in volume. The rate of the reaction is dependent upon time and temperature, and upon the degree of departure from phase equilibrium. The reaction also may be sensitive to applied stress, the application of vibrational energy, and the level of impurities in the alloy. Data from such a change are given for the steels listed in Table 6 and in Figures 32-39.

(3) A metal or alloy that undergoes a transformation from one allotropic form to another will change in volume. The change may be positive or negative, depending upon the relative specific volumes of the two phases. In steel, for example, the transformation from austenite to martensite results in a volume increase, the magnitude of which is dependent upon alloy composition.

(4) Combinations of the several mechanisms described above may occur concurrently. For example, a steel may exhibit simultaneously a positive volume change from the transformation of retained austenite and a negative volume change from the tempering of martensite. Thus, the net volume change may be positive, negative, or zero; it also may change from one to the other over a period of time as one mechanism becomes dominant over another. An example of this is shown in Figure 40, which illustrates the dimensional changes occuring in a maraging steel during the course of its heat treatment.

Additional data and recommended stabilization treatments are given in Tables 7-13 and in Figures 41-56 for 2024, 6061, and 356-T6 aluminum alloys, 310, 410, 420, 17-7PH and 17-4PH stainless steels, cast magnesium alloys, and other materials.

Residual-Stress Mechanisms

Shape distortions introduced by the relaxation of residual stresses are somewhat more difficult to analyze. Residual stresses most frequently are introduced during fabrication or heat treatment. Figures 52-56 give examples of dimensional instability in steel, as caused by previous

* Tables begin on page 27.

plastic deformation. Of particular importance in this respect are the residual stresses resulting from machining. Bonfield, et al,[2] show stresses as high as 39,000 psi (Figure 57) to exist near the surface of turned beryllium. This acquires added significance when compared to a PEL for the same material of near 2,000 psi. That such stresses are not unique to beryllium is clear from a comprehensive study by Zlatin, et al,[3] who investigated the effects of machining variables on residual stresses in a 250-grade maraging steel, Ti-8Al-1Mo-1V, Inconel 718, and Waspaloy. These residual stresses (examples are given in Figures 58-61) concentrated near the surface, can contribute significantly to dimensional instability if they are sufficiently high to induce microcreep. It is often difficult to remove these stresses, even after a high-temperature anneal (Figure 62).

Just how much such stresses can affect the dimensions of a part is illustrated by an experiment of Eggert's.[4] Using commercially obtained gage blocks of 52100 steel, Eggert electropolished off about 0.010 inch from one side. As shown in Figure 63, two blocks tested decreased 177 and 223 microinches per inch, respectively. Thus, under such conditions of residual stress, any subsequent wear or annealing or shock must be expected to induce a change in dimensions.

A different form of residual stress can be set up as a result of changes in temperature if the material in question is anisotropic in its thermal expansion. In most noncubic materials, the thermal expansion coefficient differs appreciably in different lattice directions. Several examples are shown in Figures 64 and 65. Therefore, when the temperature of a polycrystalline aggregate changes, an appreciable amount of stress can be built up between adjacent grains. Davidenkov, et al,[5] have calculated how these stresses depend upon the temperature change for a variety of materials. Their data are shown in Table 14. Figure 66 shows the extent to which such stresses can produce metallographically observable damage, even in mildly anisotropic material such as zirconium. Figures 67-70 show the extent to which dimensions and density can be affected by thermal cycling in such materials. Curing such a situation can become complex, for the thermal expansion coefficient itself is apparently influenced by both heat treatment (see Figures 71 and 72) and by the method of fabrication (see Figure 73).

Data which appear to be related to both residual-stress patterns and incomplete transformation in 52100 steel are shown in Figure 74 and in Tables 15-19. The data show not only a remarkable effect of neutron irradiation on the dimensional stability of 52100 steel, but an even more remarkable influence of the stabilization treatment on the effect of the neutron irradiation.

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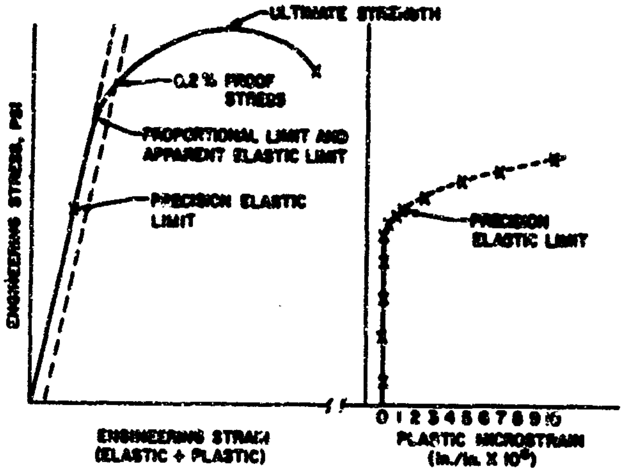


FIGURE 1. A TYPICAL STRESS-STRAIN CURVE(6)

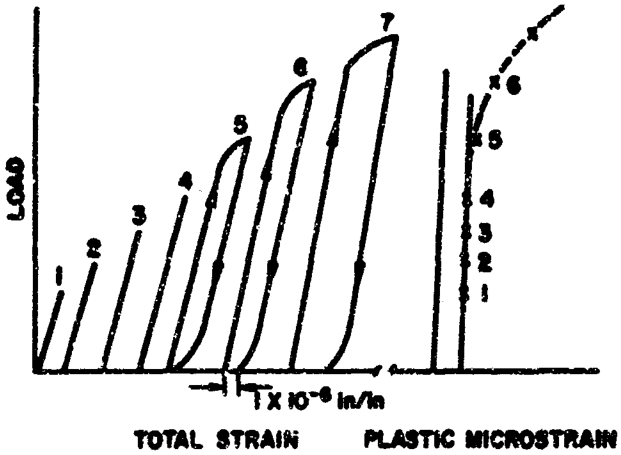


FIGURE 2. A TYPICAL PRECISION-ELASTIC-LIMIT (PEL) DETERMINATION(6)

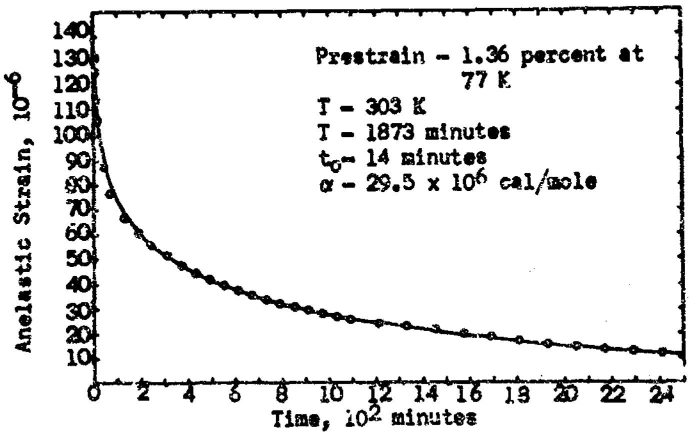


FIGURE 4. THE ELASTIC AFTER EFFECT IN A PRESTRAINED ZIRCONIUM SPECIMEN STRESSED TO 15,000 PSI AND QUICKLY UNLOADED.(7)

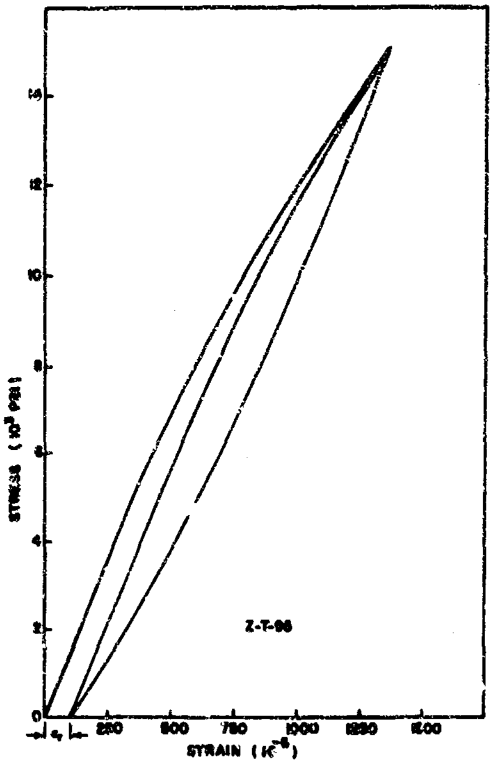


FIGURE 3. TYPICAL TWO-CYCLE ROOM-TEMPERATURE TENSILE STRESS-STRAIN DIAGRAM FOR A PRESTRAINED ZIRCONIUM SPECIMEN (0.65 % PRESTRAIN AT 77 K BY ROLLING)(7)

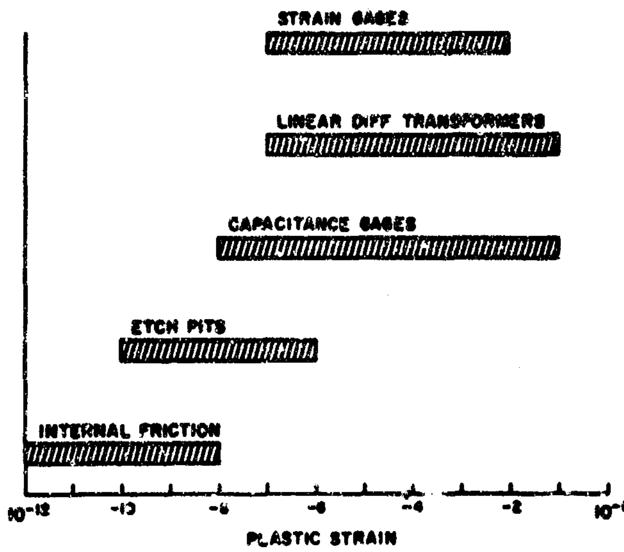


FIGURE 5. RANGES OF STRAIN MEASURED BY VARIOUS TECHNIQUES(6)

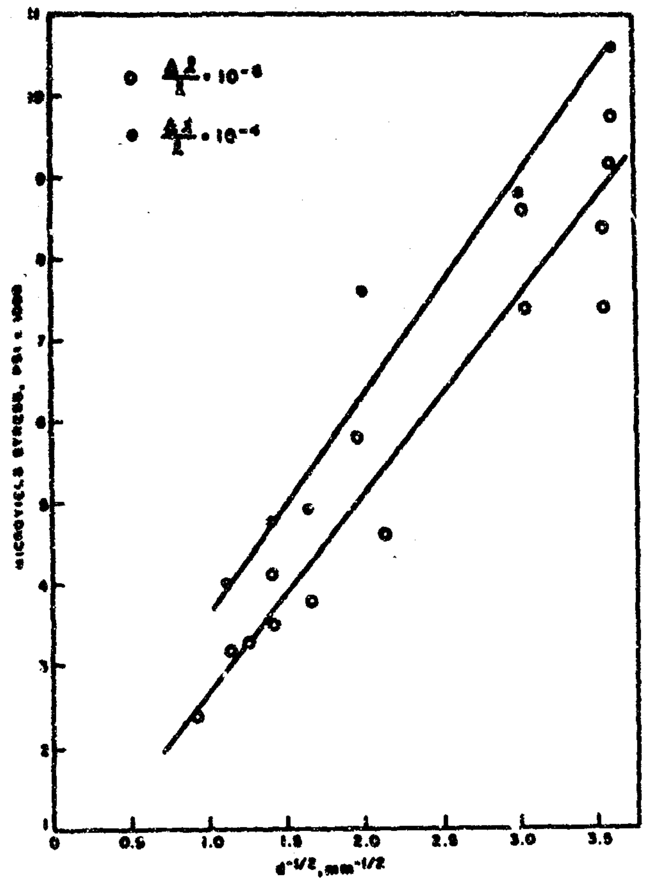


FIGURE 6. MICROYIELD STRESS OF NICKEL AS A FUNCTION OF THE GRAIN-SIZE PARAMETER, $d^{-1/2}$(8)

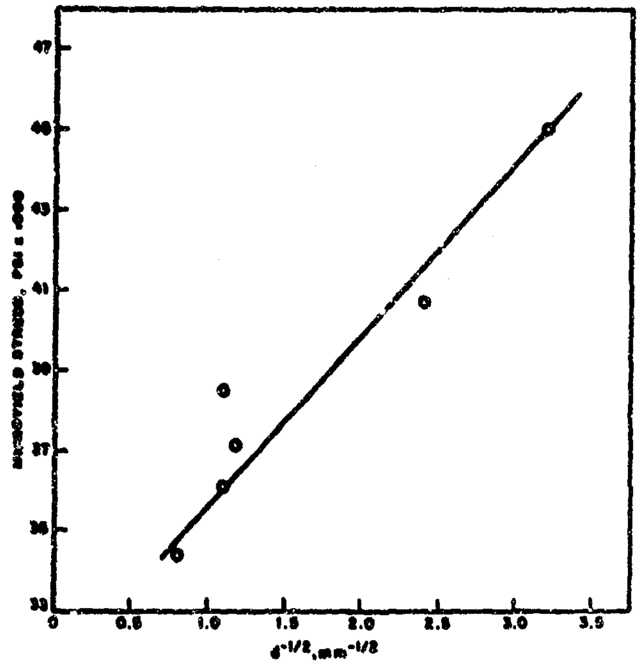


FIGURE 8. MICROYIELD STRESS OF Fe-3%Si AS A FUNCTION OF THE GRAIN-SIZE PARAMETER, $d^{-1/2}$(8)

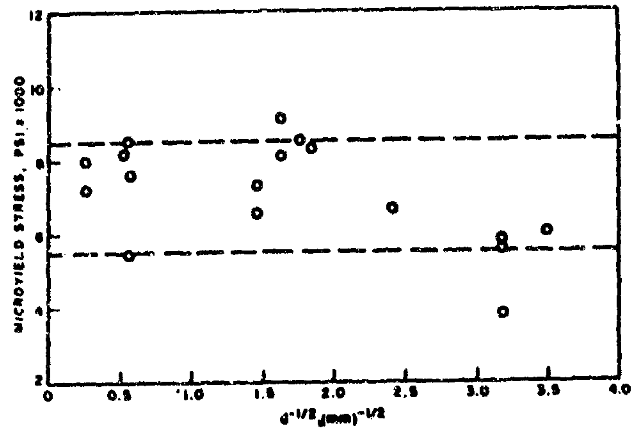


FIGURE 7. DEPENDENCE OF THE MICROYIELD STRESS OF IRON ON THE GRAIN-SIZE PARAMETER, $d^{-1/2}$(8)

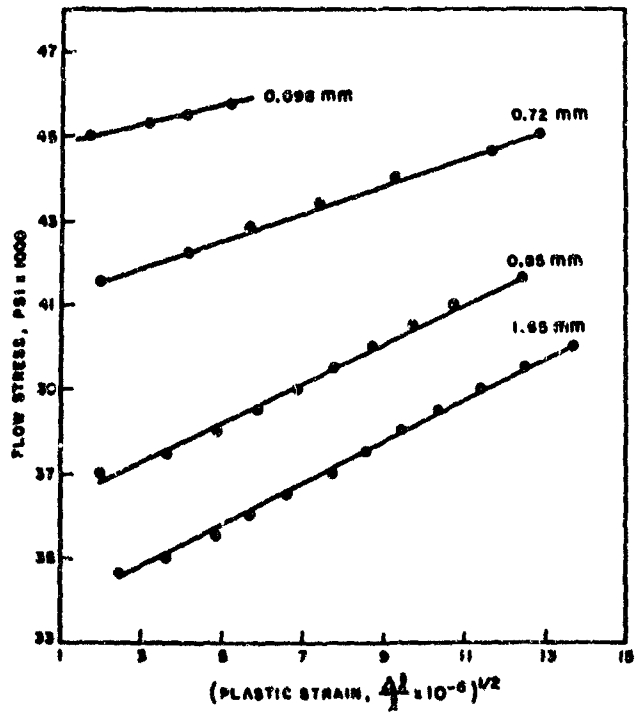


FIGURE 9. STRESS-MICROSTRAIN CURVES FOR Fe-3%Si ALLOYS OF VARIOUS GRAIN SIZES (DIAMETER IN MM) PLOTTED ON A PARABOLIC SCALE(8)

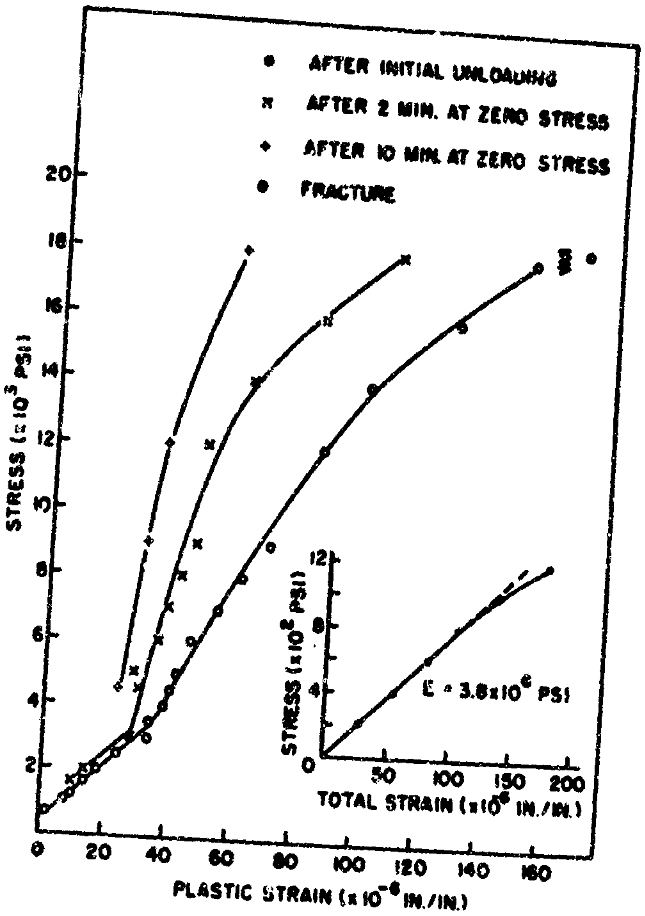


FIGURE 10. PLASTIC DEFORMATION OF BONE(9)

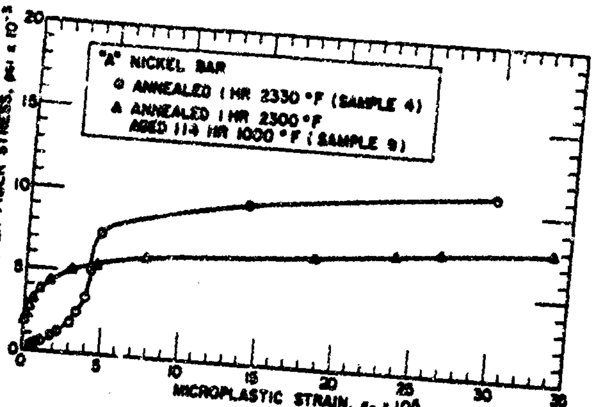


FIGURE 12. STRESS-MICROPLASTIC STRAIN CURVES FOR TWO COARSE-GRAINED ROD SAMPLES(10)

Open O's represent as-annealed behavior and open Δ's the behavior subsequent to aging 114 hours at 1000 F.

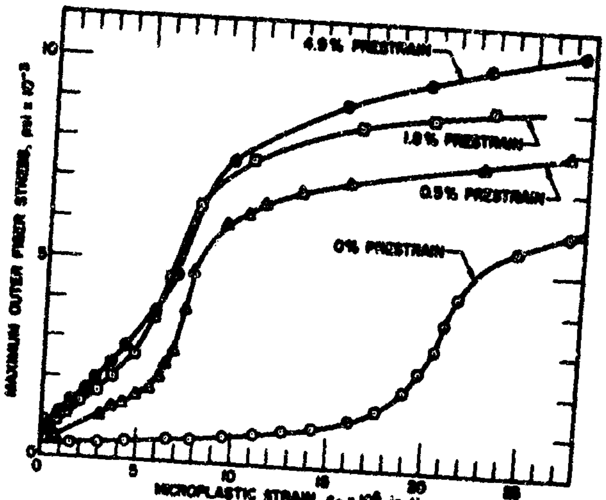


FIGURE 13. STRESS-MICROPLASTIC STRAIN CURVES FOR PRESTRAIN POLYGONIZED SAMPLES(10)

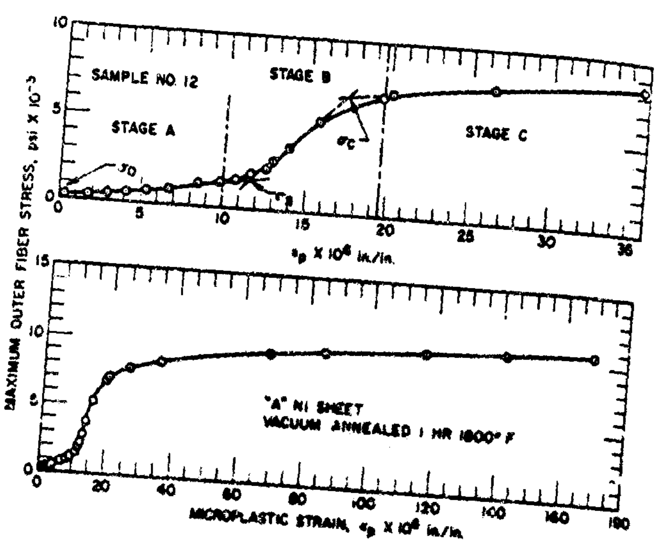


FIGURE 11. STRESS-MICROPLASTIC STRAIN CURVE FOR SHEET SAMPLE(10)

(a) Expanded plot of initial region illustrating three-stage behavior.
(b) Plot of total plastic deformation of this sample.

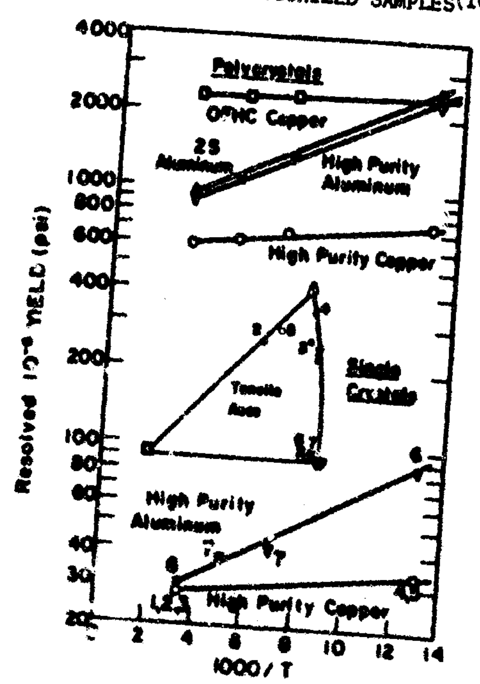


FIGURE 14. 10^{-6} YIELD OF COPPER AND ALUMINUM SINGLE CRYSTALS AND POLYCRYSTALLINE MATERIALS(11)

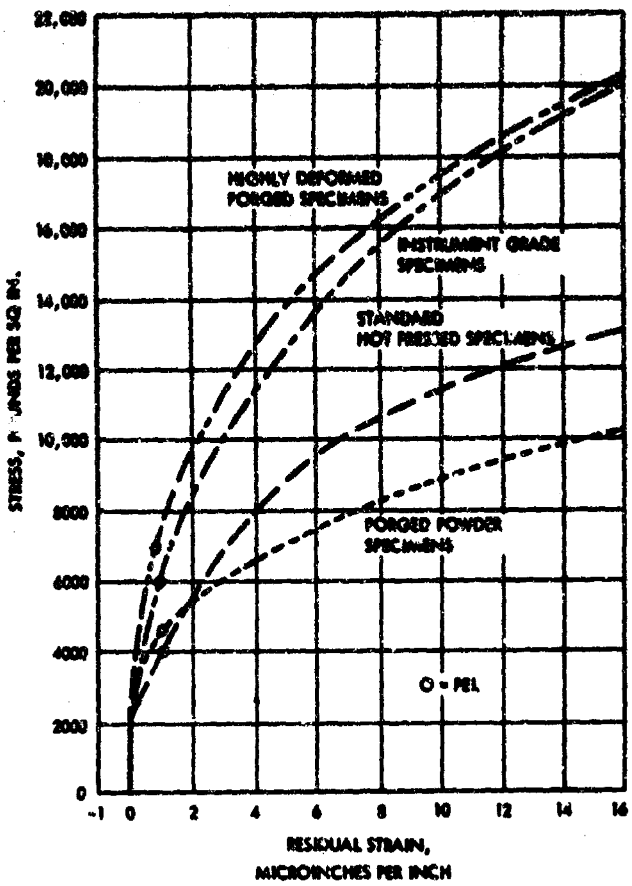


FIGURE 15. PEL COMPARISONS FOR VARIOUS TYPES OF BERYLLIUM[12]

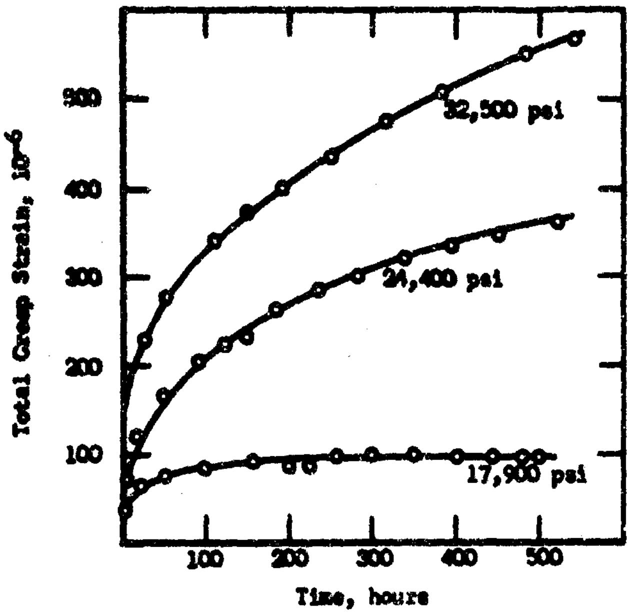


FIGURE 17. TOTAL CREEP CURVES OF 310 STAINLESS STEEL AT 85 F[13]

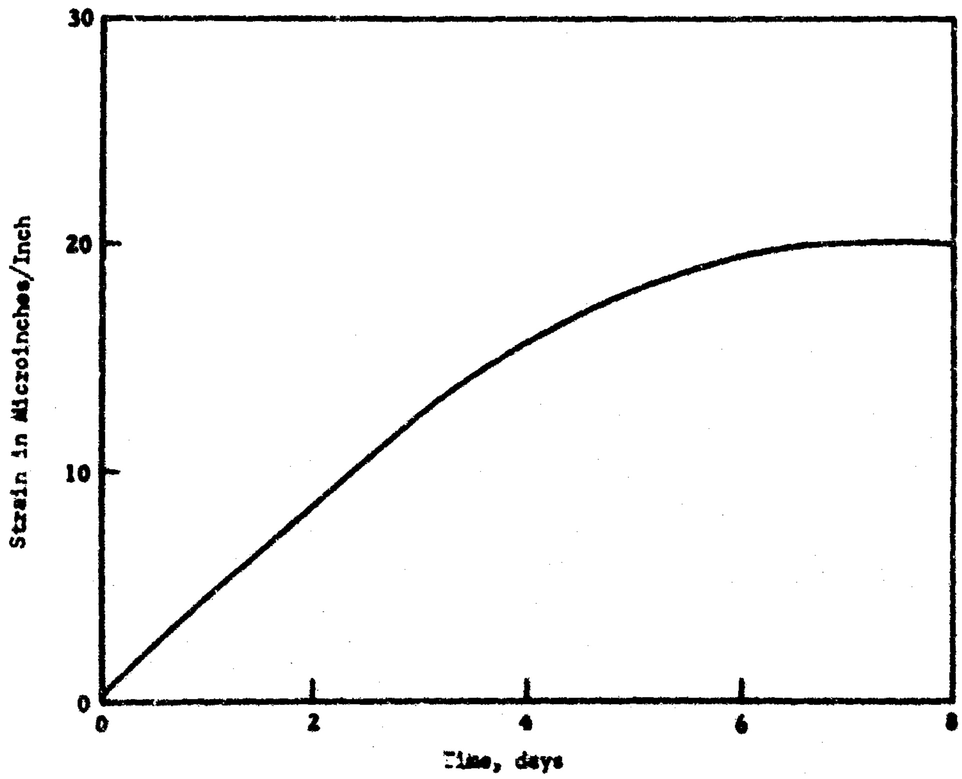


FIGURE 16. HOT-PRESSED BERYLLIUM STRESSED-DIMENSIONAL-STABILITY MICROCREEP AT 12,000 PSI APPLIED STRESS[18]

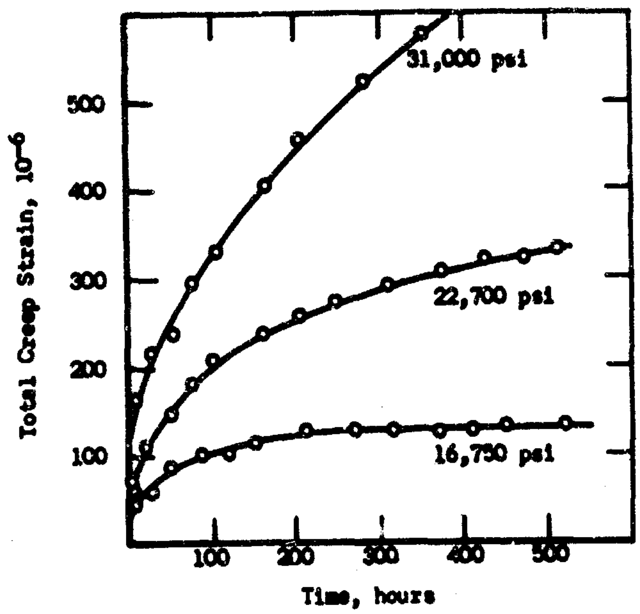


FIGURE 18. TOTAL CREEP CURVES OF 310 STAINLESS STEEL AT 150 F(13)

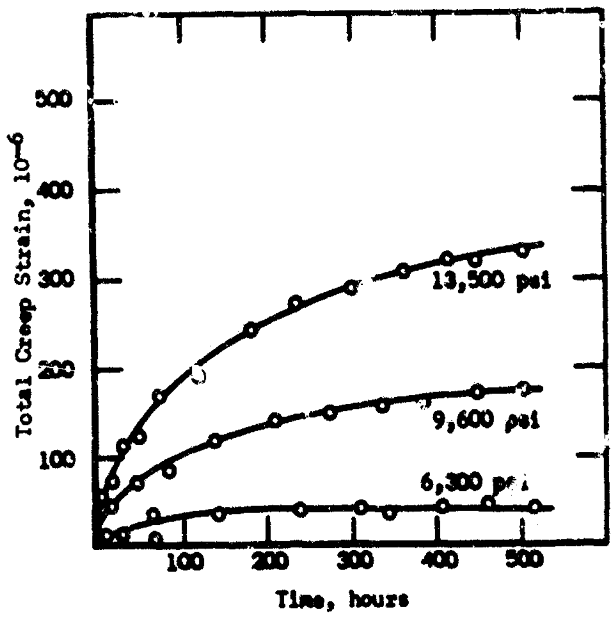


FIGURE 20. TOTAL CREEP CURVES OF 356-T6 ALUMINUM AT 85 F(13)

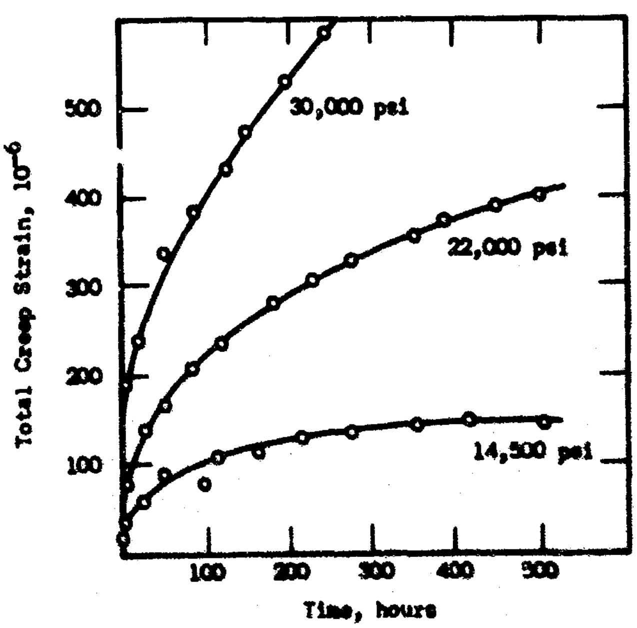


FIGURE 19. TOTAL CREEP CURVES OF 310 STAINLESS STEEL AT 200 F(13)

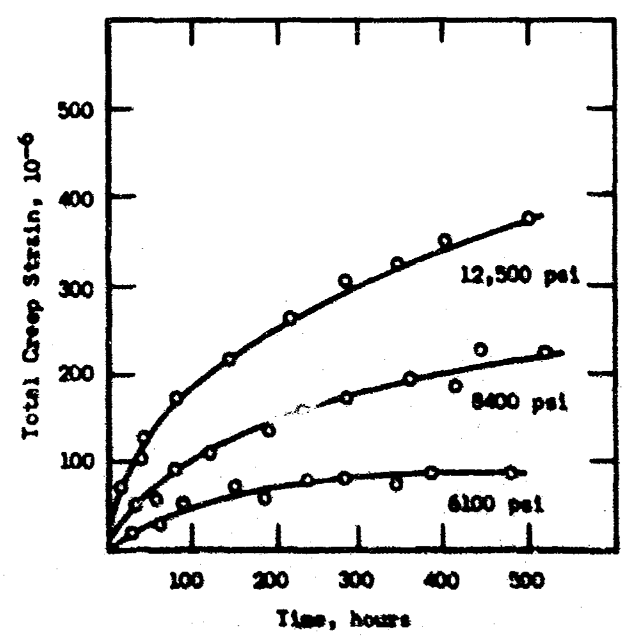


FIGURE 21. TOTAL CREEP CURVES OF 356-T6 ALUMINUM AT 150 F(13)

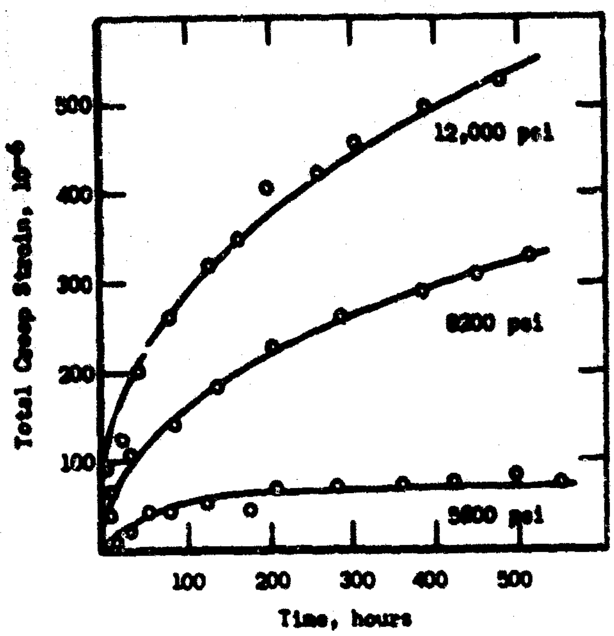


FIGURE 22. TOTAL CREEP CURVES OF 356-T6 ALUMINUM AT 200 F(13)

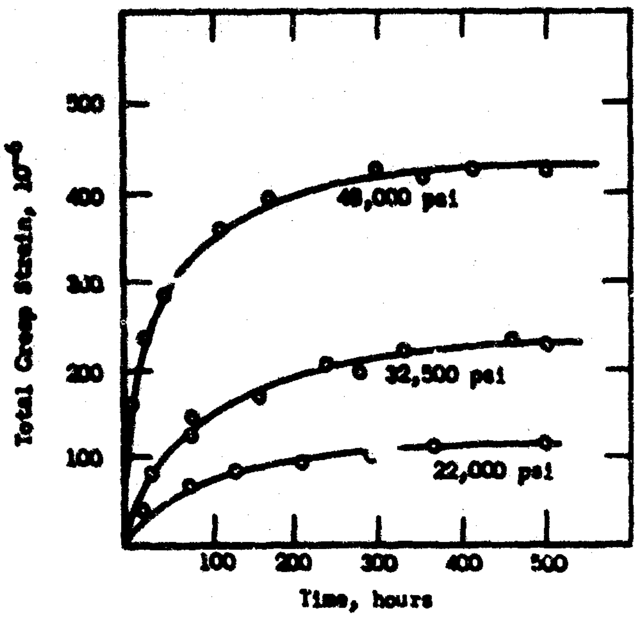


FIGURE 24. TOTAL CREEP CURVES OF INVAR 36 AT 150 F(13)

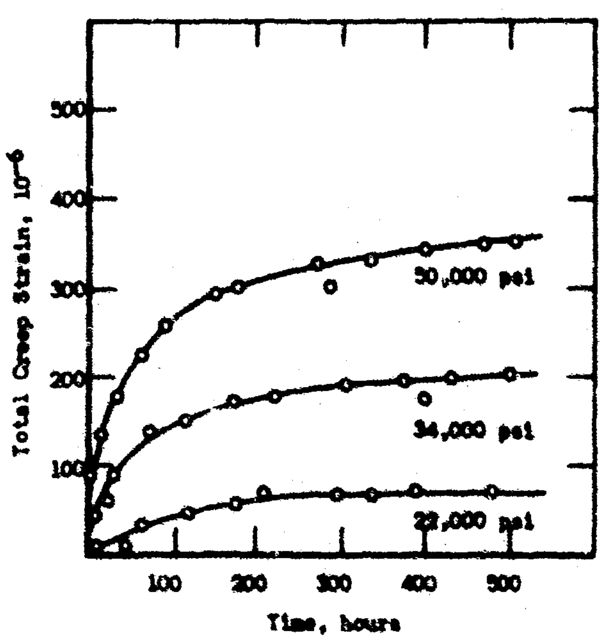


FIGURE 23. TOTAL CREEP CURVES OF INVAR 36 AT 85 F(13)

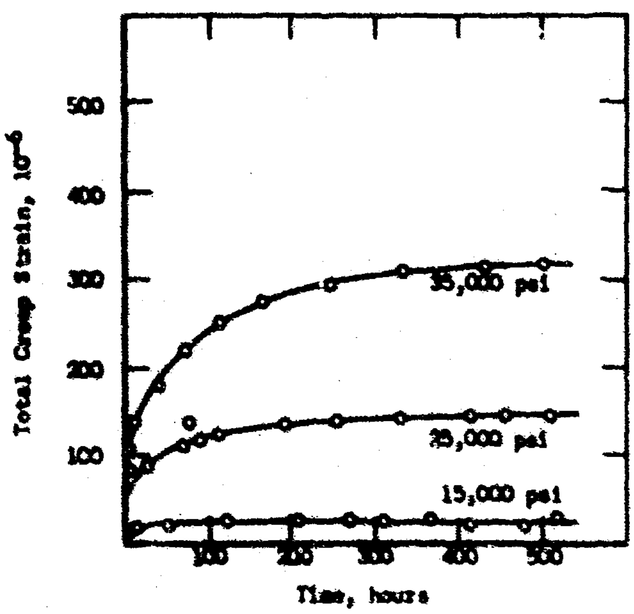


FIGURE 25. TOTAL CREEP CURVES OF INVAR 36 AT 200 F(13)

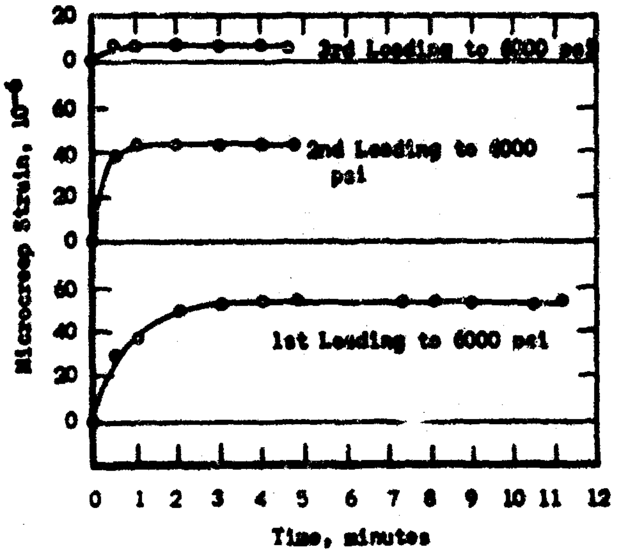


FIGURE 26. REPEATED ROOM-TEMPERATURE CREEP TESTS OF HIGH-PURITY COPPER(14)

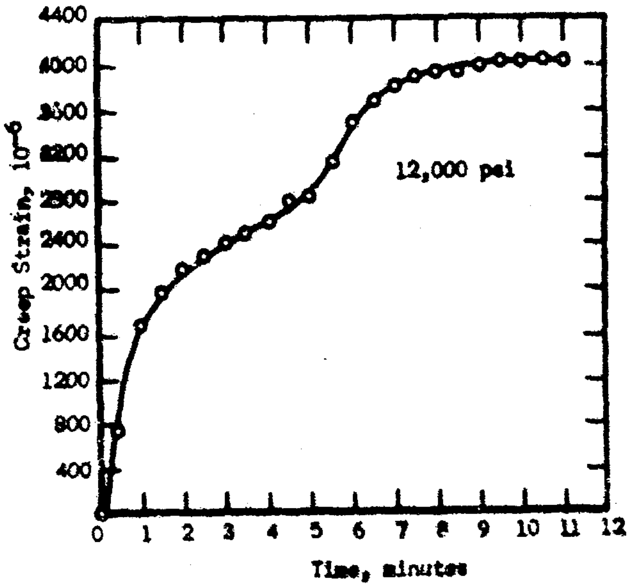


FIGURE 27. CREEP OF HIGH-PURITY ALUMINUM AT -196 C(14)

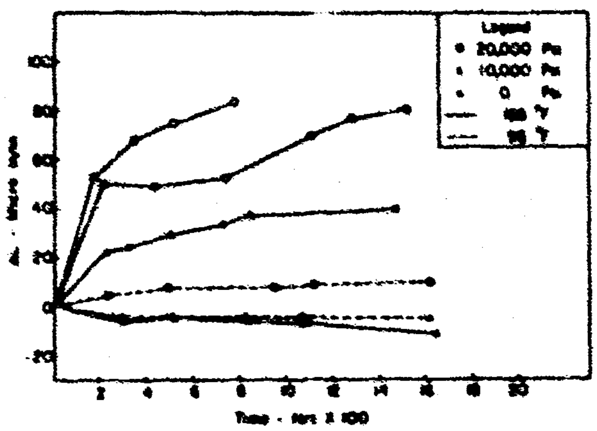


FIGURE 28. DIMENSIONAL STABILITY OF STRUCTURE C - SUBCOOLED 52100 STEEL - AT VARIOUS STRESSES AND TEMPERATURES(15)

Austenitize 1/2 hr at 1550 F, O.Q. at 90 F + Temper 1/2 hr at -321 F, AC + 1/2 hr at 250 F, AC, repeated 10 times.

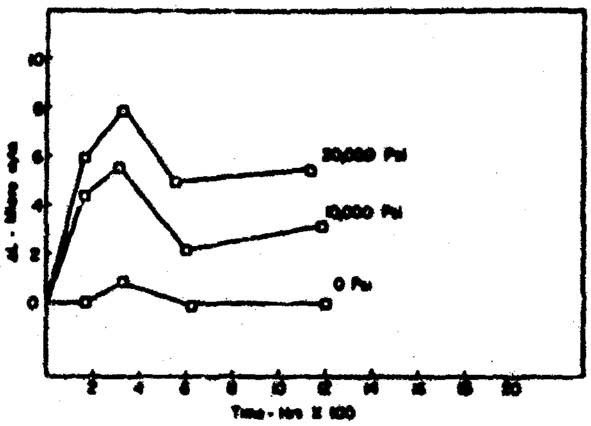


FIGURE 29. DIMENSIONAL STABILITY OF STRUCTURE D - AUSTEMPERED 52100 STEEL - AT VARIOUS STRESSES AND 160 F(15)

Austenitize 1/2 hr at 1550 F, Austemper 1 hr at 500 F, AC.

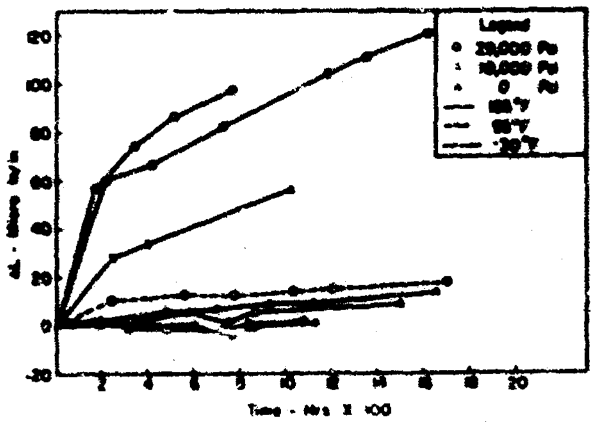


FIGURE 30. DIMENSIONAL STABILITY OF STRUCTURE A - TEMPERED 10 HR AT 250 F AT VARIOUS STRESSES AND TEMPERATURES(15)

Austenitize 1/2 hr at 1550 F, O.Q. at 90 F + Temper 10 hr at 250 F.

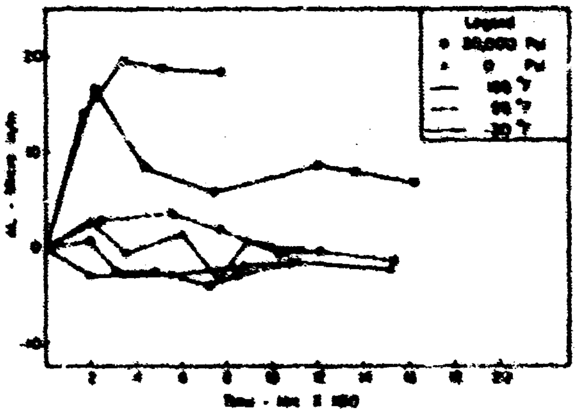


FIGURE 31. DIMENSIONAL STABILITY OF STRUCTURE B - TEMPERED 1 HR AT 500 F AT VARIOUS STRESSES AND TEMPERATURES(15)

Austenitize 1/2 hr at 1550 F, O.Q. at 90 F + Temper 1 hr at 500 F.

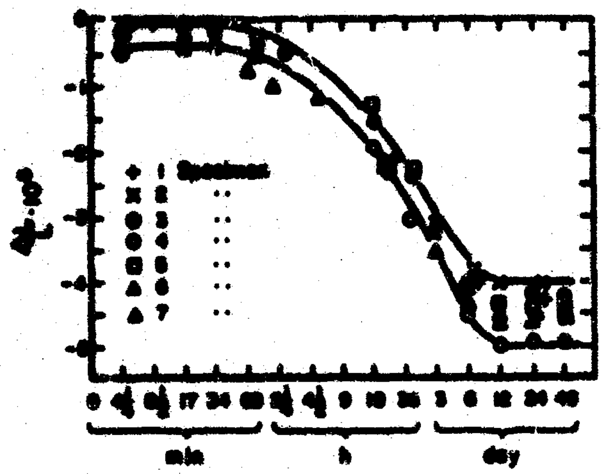


FIGURE 32. DIMENSIONAL CHANGE OF STEEL 1 DURING QUENCH AGING AT 75 C. PREVIOUS TREATMENT: WATER QUENCHED FROM 710 C(16)

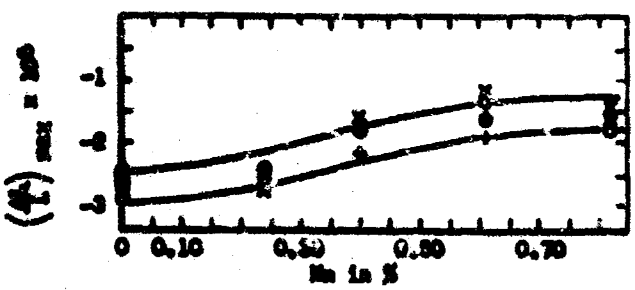


FIGURE 35. DEPENDENCE OF MAXIMUM DIMENSIONAL CHANGES ON MANGANESE CONTENT. QUENCHED FROM 600 C, AGED AT 75 C(16)

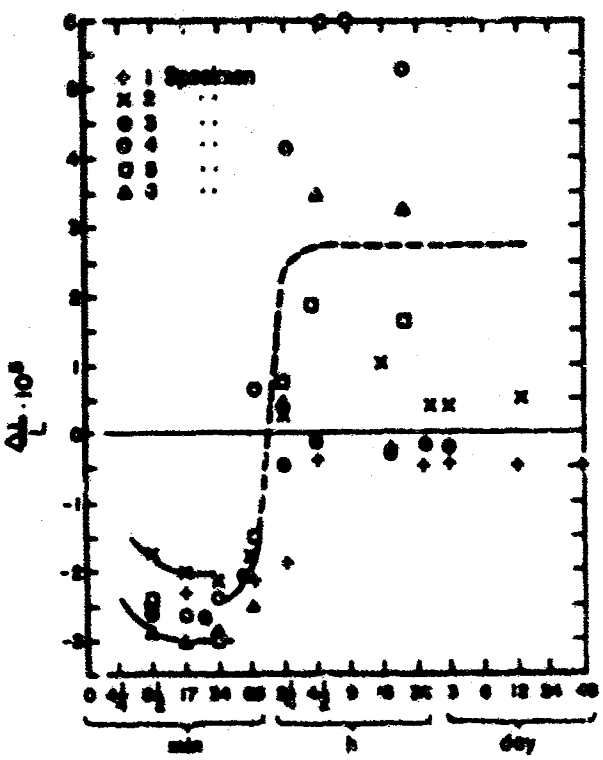


FIGURE 33. DIMENSIONAL CHANGE OF STEEL 2 DURING QUENCH AGING AT 200 C. PREVIOUS TREATMENT: WATER QUENCHED FROM 710 C(16)

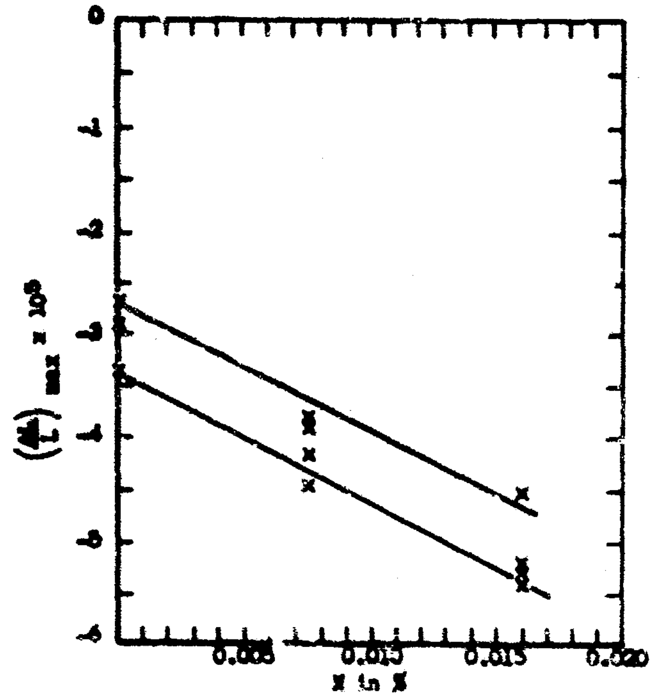


FIGURE 36. DEPENDENCE OF MAXIMUM DIMENSIONAL CHANGES ON NITROGEN CONTENT. QUENCHED FROM 650 C, AGED AT 75 C(16)

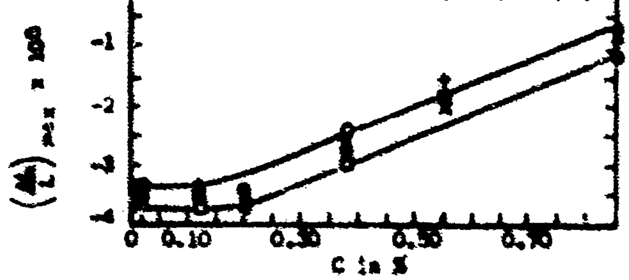


FIGURE 34. DEPENDENCE OF MAXIMUM DIMENSIONAL CHANGES ON CARBON CONTENT. QUENCHED FROM 650 C. AGED AT 75 C(16)

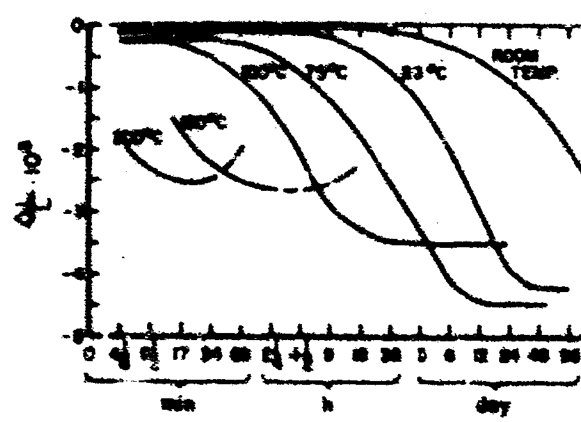


FIGURE 37. DIMENSIONAL CHANGE OF STEEL 1 DURING QUENCH AGING AT DIFFERENT TEMPERATURES AFTER QUENCH FROM 710 C(16)

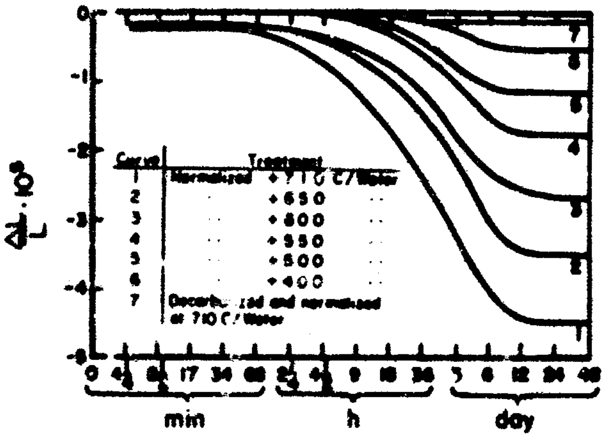


FIGURE 38. DIMENSIONAL CHANGE OF STEEL 1 AT 75 C AFTER QUENCHING FROM VARIOUS TEMPERATURES(16)

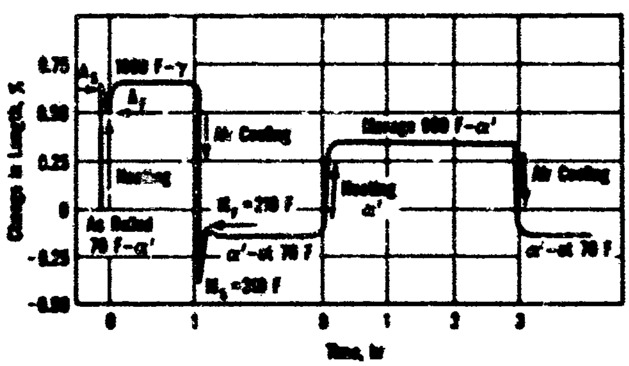


FIGURE 40. SUMMARY OF MARAGING TREATMENT IN ITS ENTIRETY(17)

(a) Change in length of specimen plotted as a function of the time of the heat treatment.

(b) The temperatures and the phases which are stable at that temperature are indicated.

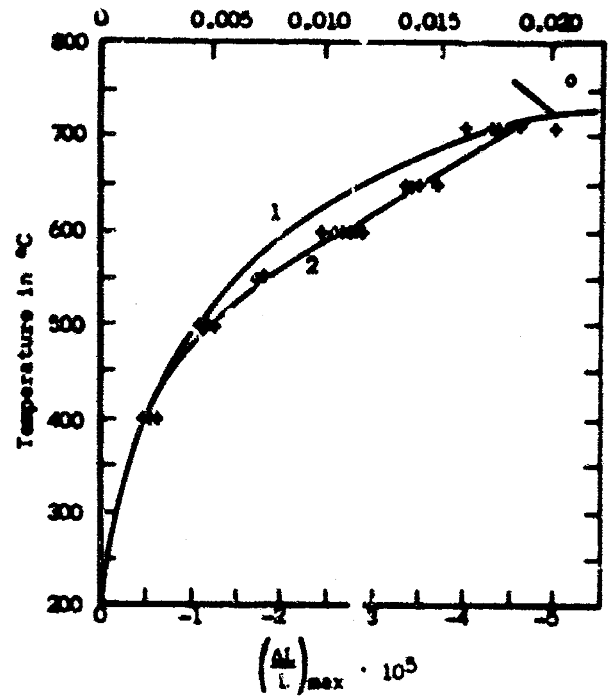


FIGURE 39. CORRELATION BETWEEN MAXIMUM DIMENSIONAL CHANGES AND CARBON IN SOLUTION IN ALPHA-IRON. CURVE 1: CARBON SOLUBILITY FROM THE EQUILIBRIUM DIAGRAM. CURVE 2 DIMENSIONAL CHANGES (AGED AT 75 C) AS A FUNCTION OF QUENCHING TEMPERATURE(16)

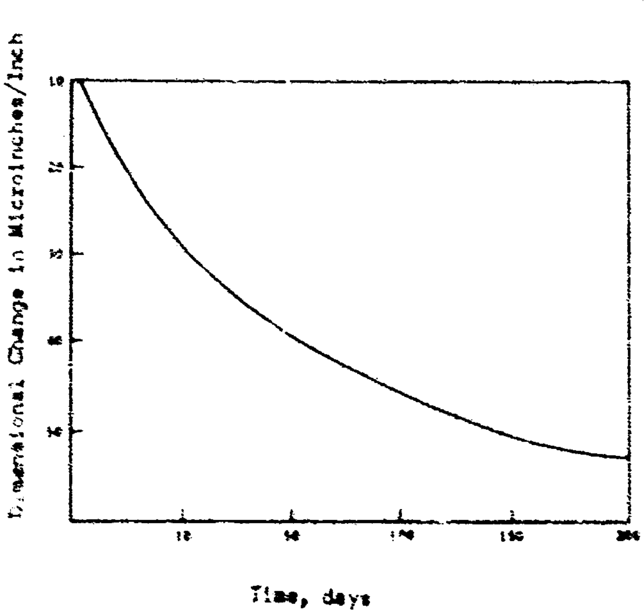


FIGURE 41. 2024 ALUMINUM ALLOY UNSTRESSED DIMENSIONAL STABILITY, NATURAL AGING AFTER SOLUTION HEAT TREATMENT(18)

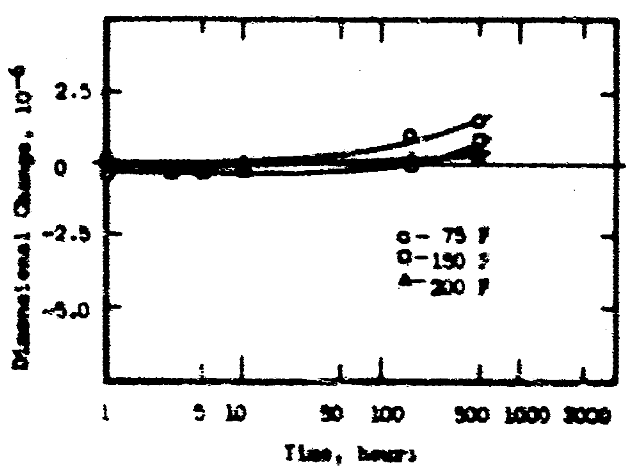


FIGURE 42. DIMENSIONAL CHANGES OF 6061 ALUMINUM STORED AT 75, 150, AND 200 F(13)

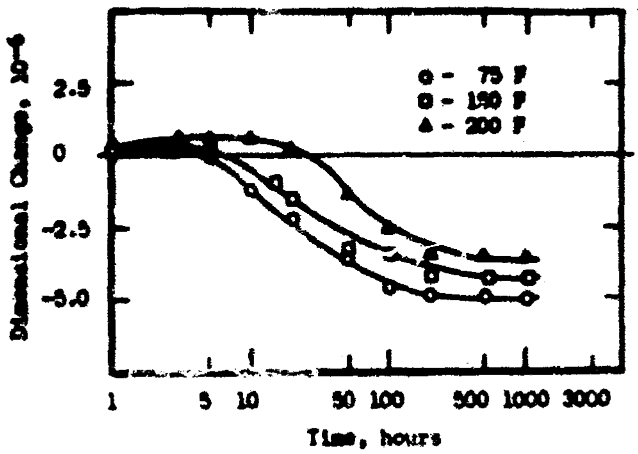


FIGURE 43. DIMENSIONAL CHANGES OF 356-T6 ALUMINUM STORED AT 75, 150, AND 200 F(13)

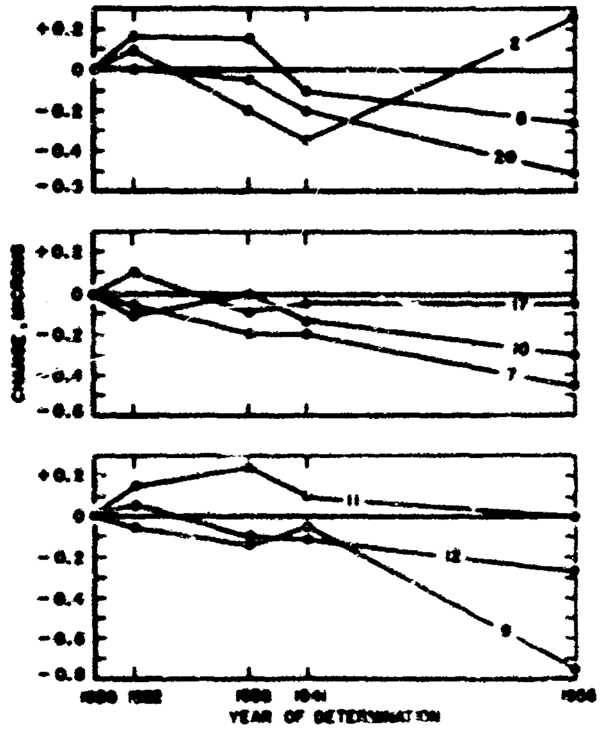


FIGURE 46. CHANGE IN LENGTH OF IRON-CHROMIUM DECIMETER BARS(19)

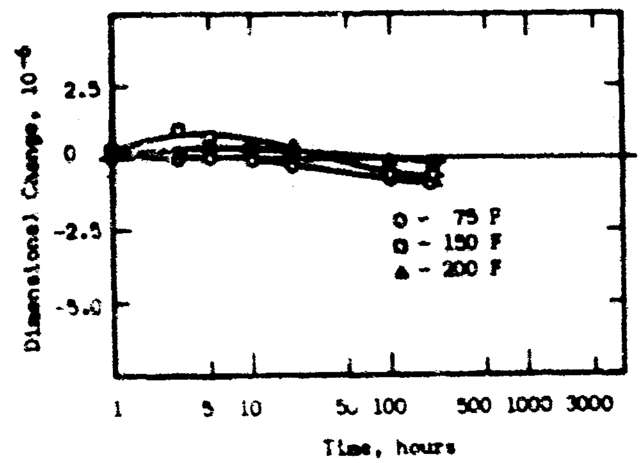


FIGURE 44. DIMENSIONAL CHANGES OF 310 STAINLESS STEEL STORED AT 75, 150, AND 200 F(13)

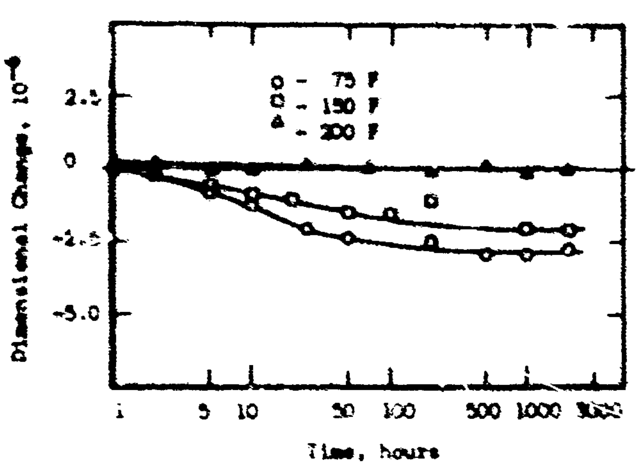


FIGURE 45. DIMENSIONAL CHANGES OF FREE-CUT INVAR 36 STORED AT 75, 150, AND 200 F(13)

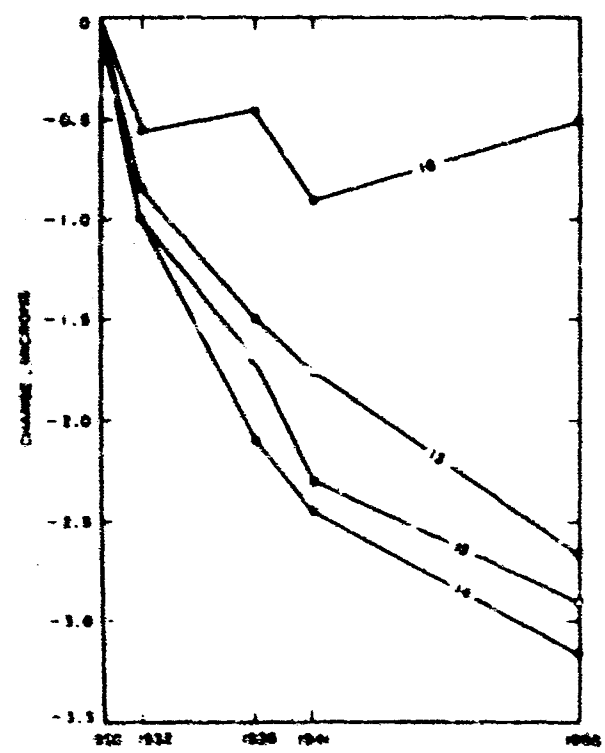


FIGURE 47. CHANGE IN LENGTH OF IRON-CHROMIUM DECIMETER BARS(19)

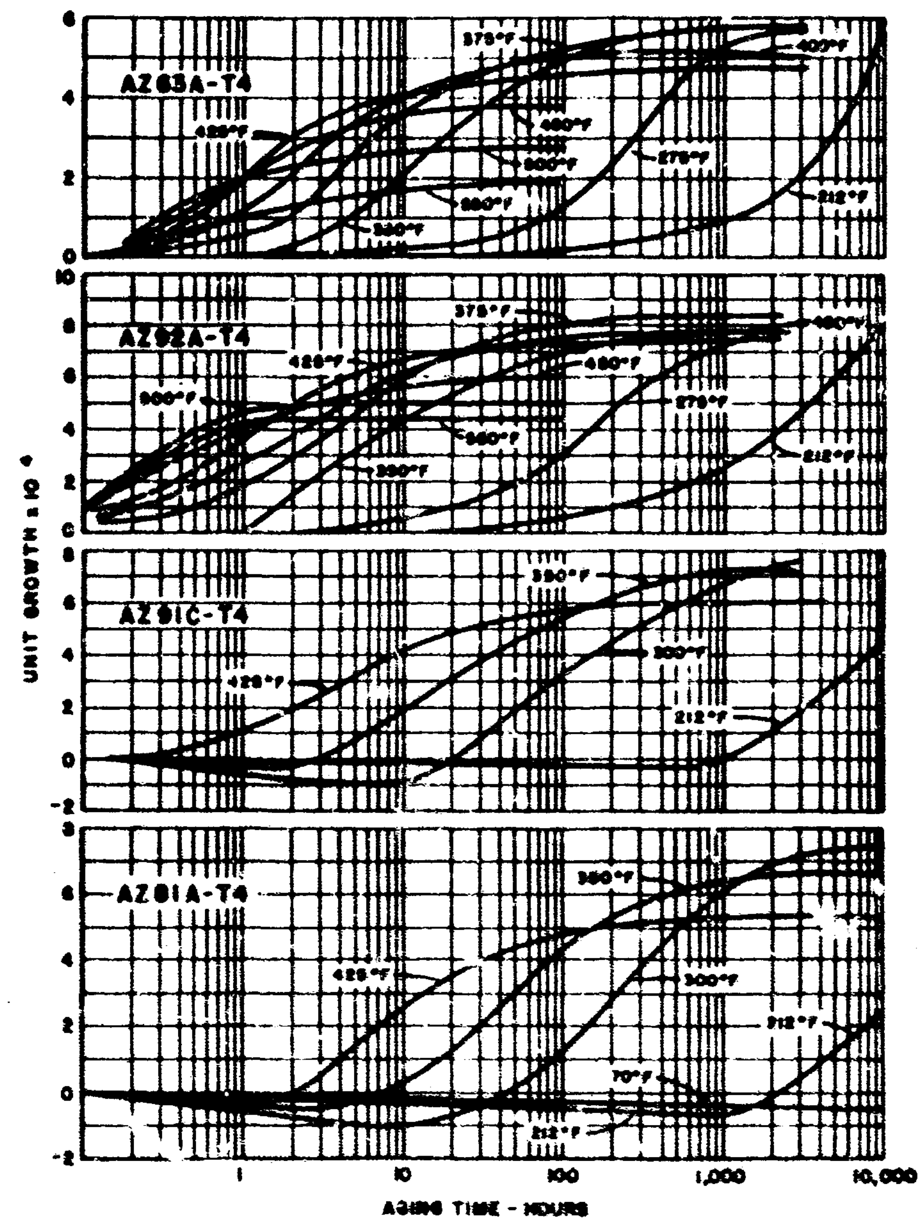


FIGURE 48. GROWTH RATES FOR MAGNESIUM CASTING ALLOYS OF THE Mg-Al-Zn SYSTEM[20]

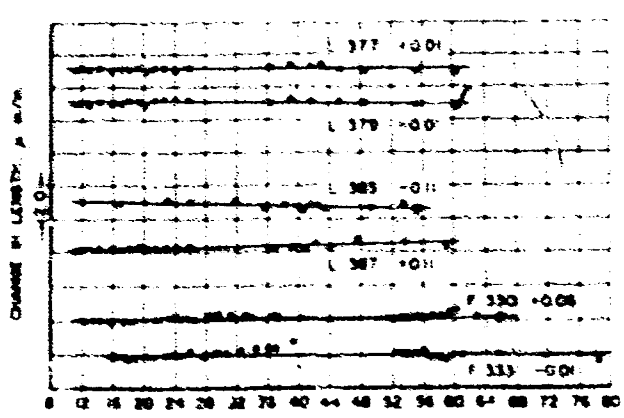


FIGURE 49. DIMENSIONAL STABILITY OF STEELS WITH ANNEALED CORES AND CASE-HARDENED SURFACES[21]

L 377 and L 379 are 1010 steel, packed carburized; L 383 and L 387 are 1010 steel, carbonitrided; F 330 and F 333 are 410 stainless steel, nitrided. The numbers preceded by + or - indicate average stability in (microin./in.)/yr. (See note on page 20.)

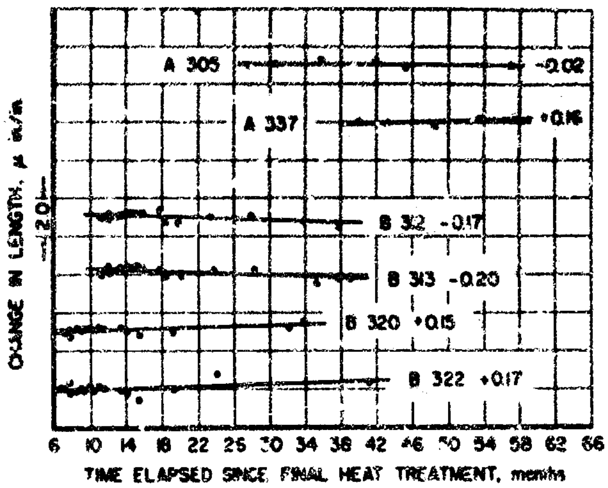


FIGURE 50. DIMENSIONAL STABILITY OF STEELS WITH ANNEALED CORES AND NITRIDED SURFACES[21]

A represents 304 stainless; B, Type 405 stainless steel. The numbers preceded by + or - indicate average stability in (microin./in.)/yr. (See note.)

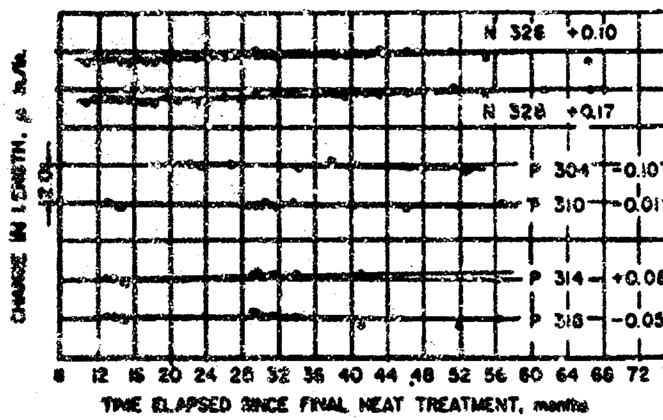


FIGURE 51. DIMENSIONAL STABILITY OF CASE-HARDENED STEELS WITH MEDIUM-HARD CORES[21]

N represents Nitralloy 135 modified; P, Type 17-4PH stainless steel. The numbers preceded by + or - indicate average stability in (microin./in.)/yr. (See note.)

NOTE TO FIGURES 49, 50, and 51

The 1010 steel was carburized by packing in a proprietary carburizing compound and heating to 1750 F for 10 hours. The case was hardened by reheating the specimens to 1625 F for 15 minutes in a neutral salt bath and quenching in brine. The specimens were stabilized by immediately refrigerating at -140 F for 18 to 24 hours, tempering at 250 F for 1 hour, re-refrigerating at -140 F for 24 hours, and again tempering at 250 F for 9 hours.

All surfaces were ground to remove only a portion of the case. After grinding, the specimens were stress relieved at 250 F for 1-1/2 hr and any residual magnetism remaining from contact with the grinding chuck removed with a demagnetizing coil.

This steel, heat treated, stabilized, and fabricated as described, possesses qualities of ultrastability which to date have been unsurpassed. Over a four-year period the measured changes in length for two specimens have been +0.01 and -0.01 x 10^{-6} (in./in.)/yr. Curves illustrating this performance are shown in Figure 49 (specimens L 377 and L 379).

Almost as good a degree of stability was obtained by hardening the surfaces of 1010 steel specimens after carbonitriding. The specimens were carbonitrided in an appropriate atmosphere at 1600 F for 4-1/2 hours to give a case depth of about 0.022 in. The surfaces were hardened by heating the specimens to 1650 F in neutral salt and quenching in brine, followed by stabilization and tempering treatments identical to those given the carburized blocks. The hardened faces were ground to leave about 0.018 in. of case on the gaging surfaces, but all of the case on the non-gaging surfaces was completely removed. Specimens were subsequently stress relieved at 275 F for 2 hours, demagnetized, and lapped. Over a four-year period the change in length was -0.11 and +0.11 x 10^{-6} (in./in.)/yr for specimens L 383 and L 387, respectively, (see Figure 49).

Annealed 410 stainless steel was case hardened by nitriding. Nitriding was performed in an ammonia atmosphere at 1020 F for 40 to 44 hours. The case produced was 0.009 in. thick and had a hardness in excess of R_c 65. All faces were ground to remove the white layer and prepare the gaging surfaces for lapping. The specimens were than demagnetized, stress relieved at 975 F for 3 hours and lapped to size.

The nitrided 410 stainless steel specimens have been observed for about 5 years and have proven to be extremely stable. As shown in Figure 49, the instability is less than 0.1 x 10^{-6} (in./in.)/yr, being +0.06 and -0.01 x 10^{-6} for specimens F 330 and F 333, respectively.

The principle of applying nitrided cases to annealed materials was extended to types 304 and 405 stainless steel. The degree of stability exhibited by all specimens observed for periods varying from fourteen months to almost three years was an excellent 0.20 x 10^{-6} (in./in.)/yr or better (Figure 50).

Nitralloy 135 modified was treated in several ways. The best treatment consisted of hardening from 1725 F and tempering at 1200 F for 2 hours to produce a hardness of R_c 34. Specimens were nitrided at 1050 F for about 48 hours. The white layer was ground off of the nongaging surfaces and the gaging surfaces were ground in preparation for lapping. After a stress relief of 975 F for 3 hours, the specimens were rough lapped and then stress relieved a second time at 1000 F for 2 hours prior to finish lapping. The gage blocks thus produced had a high degree of stability. The two specimens observed (N 326 and N 328) showed a very slight growth rate of 0.10 and 0.17 x 10^{-6} (in./in.)/yr, respectively (Figure 51).

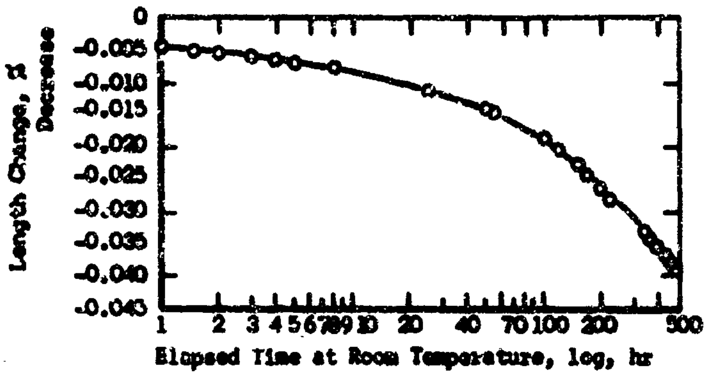


FIGURE 52. ELASTIC AFTER EFFECT FOR EUTECTOID STEEL BAR CONSISTING OF LAMELLAR PEARLITE AFTER PLASTIC ELONGATION (6%)(22)

Specimen Diameter 6 mm

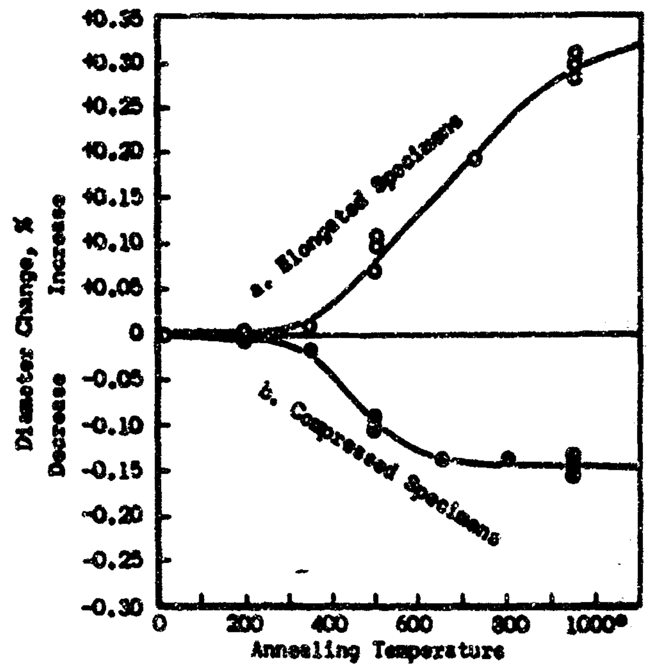


FIGURE 54. RELATION BETWEEN DIAMETER CHANGE AND ANNEALING TEMPERATURE AFTER COLD WORKING FOR EUTECTOID STEEL BARS CONSISTING OF LAMELLAR PEARLITE(22)

Holding time at annealing temp 30 min
Curve a: for 5.3% elongated specimens,
Curve b: for 5.5% compressed specimens.

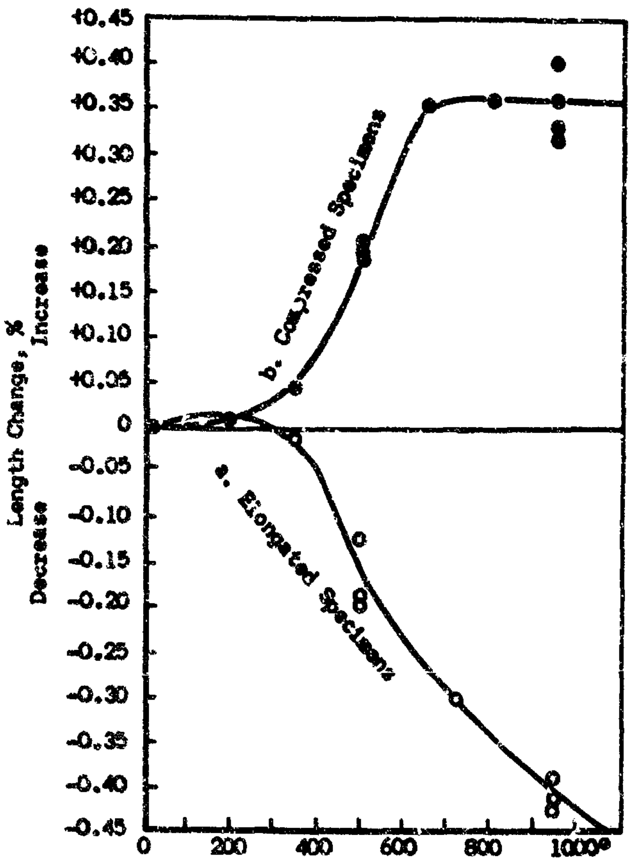


FIGURE 53. RELATION BETWEEN LENGTH-CHANGE AND ANNEALING TEMPERATURE AFTER COLD WORKING FOR EUTECTOID STEEL BARS CONSISTING OF LAMELLAR PEARLITE(22)

Curve a: for 5.3% elongated specimens,
Curve b: for 5.5% compressed specimens.

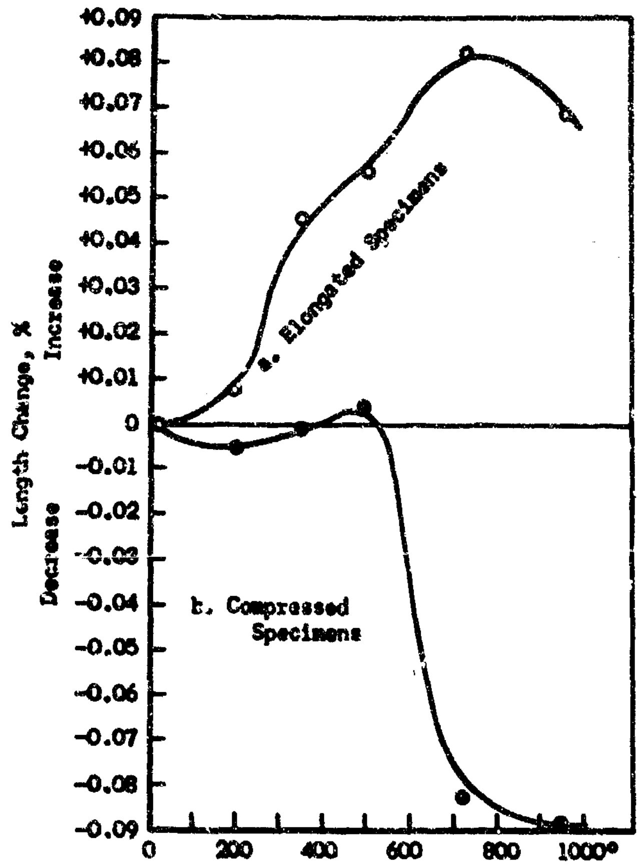


FIGURE 55. RELATION BETWEEN LENGTH CHANGE AND ANNEALING TEMPERATURE AFTER COLD WORKING FOR EUTECTOID STEEL BARS CONSISTING OF GRANULAR PEARLITE(22)

Holding time at annealing temp 30 min
Curve a: for 5.6% elongated specimens
Curve b: for 5.8% compressed specimens

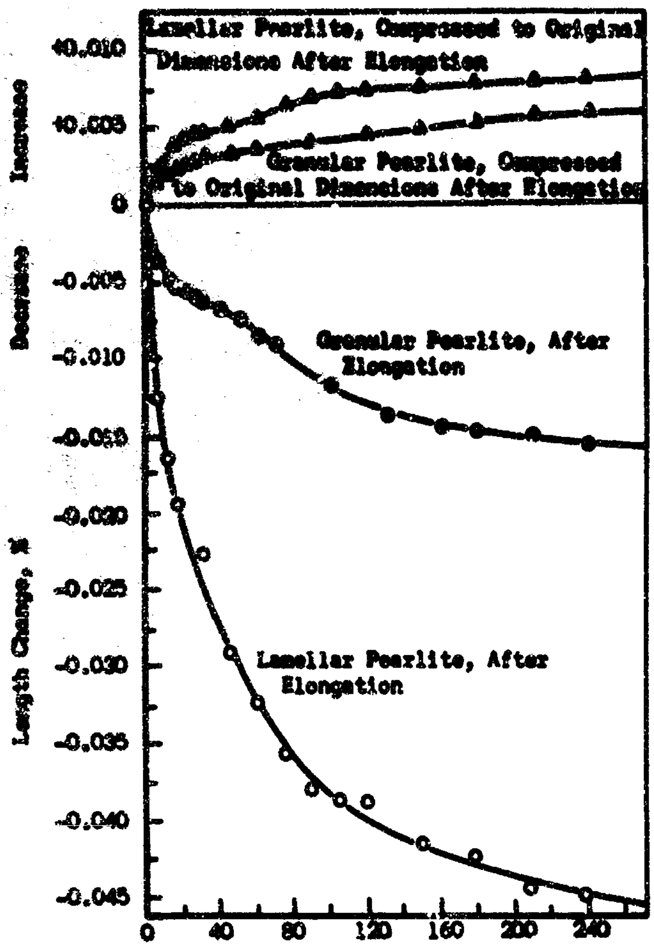


FIGURE 56. ELASTIC AFTER EFFECT OF EUTECTOID STEEL BAR AFTER ELONGATION OR COMPRESSION TO THE ORIGINAL DIMENSIONS AFTER 4% ELONGATION(22)

Room Temperature: about 10 degrees

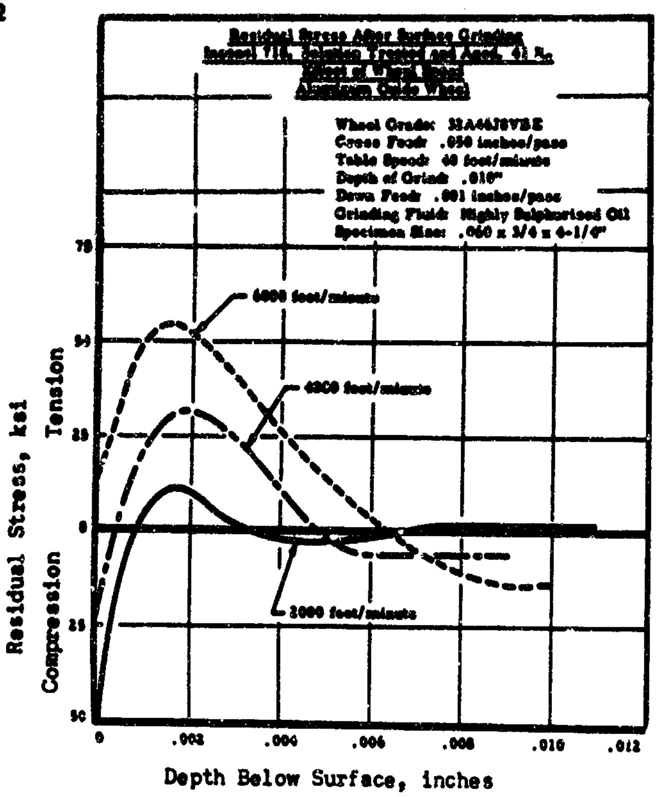


FIGURE 58. EFFECT OF WHEEL SPEED ON RESIDUAL STRESS AFTER SURFACE GRINDING INCONEL 718(3)

Solution treated and aged, 41 R_C aluminum oxide wheel

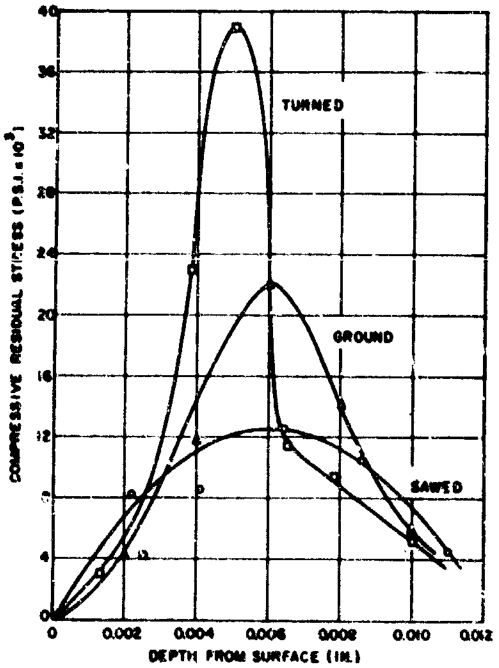


FIGURE 57. VARIATION OF RESIDUAL STRESS DISTRIBUTION WITH PREPARATION TECHNIQUE(2)

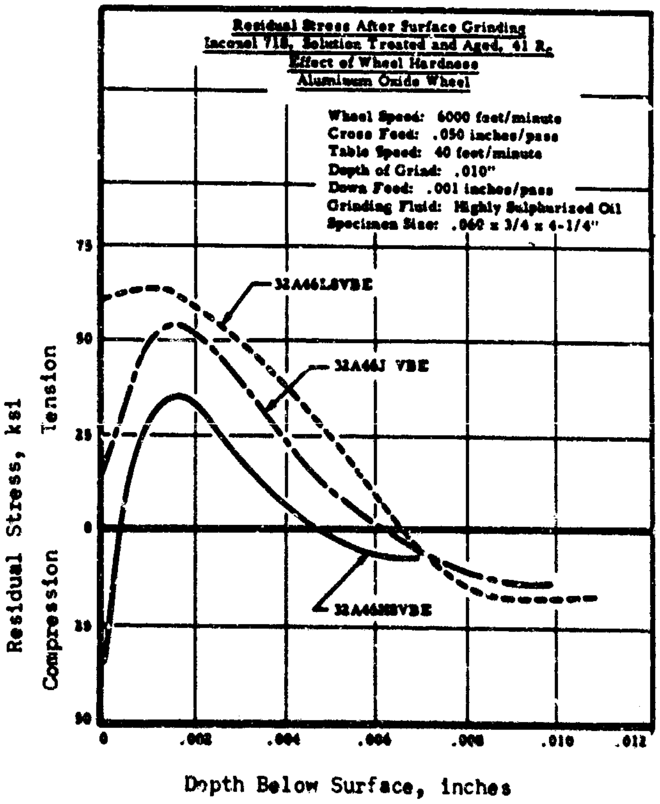


FIGURE 59. EFFECT OF WHEEL HARDNESS ON RESIDUAL STRESS AFTER SURFACE GRINDING INCONEL 718(3)

Solution treated and aged, 41 R_C aluminum oxide wheel

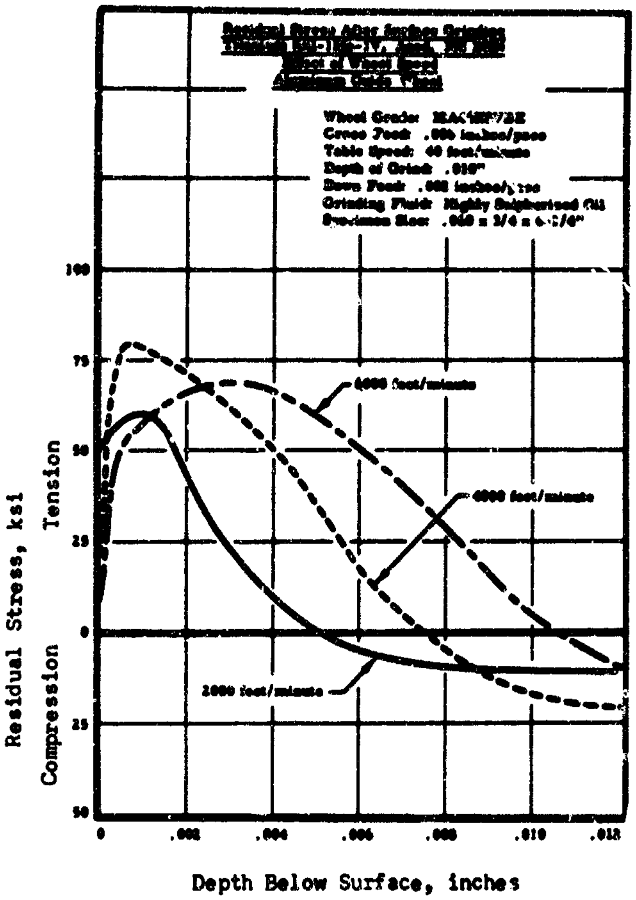


FIGURE 60. EFFECT OF WHEEL SPEED ON RESIDUAL STRESS AFTER SURFACE GRINDING TITANIUM 8Al-1Mo-1V(3)

Aged, 302 BHN aluminum oxide wheel

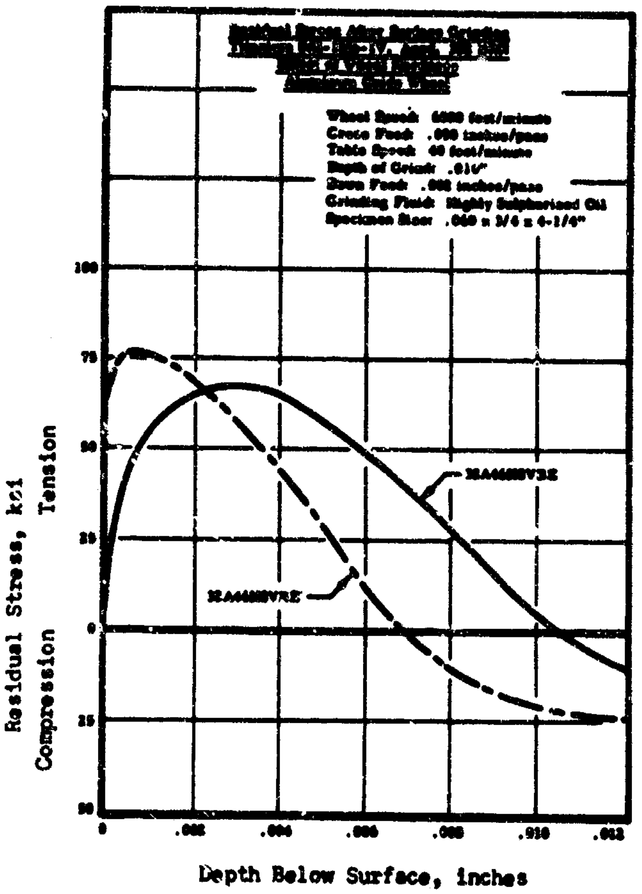


FIGURE 61. EFFECT OF WHEEL HARDNESS ON RESIDUAL STRESS AFTER SURFACE GRINDING TITANIUM 8Al-1Mo-1V(3)

Aged, 302 BHN aluminum oxide wheel

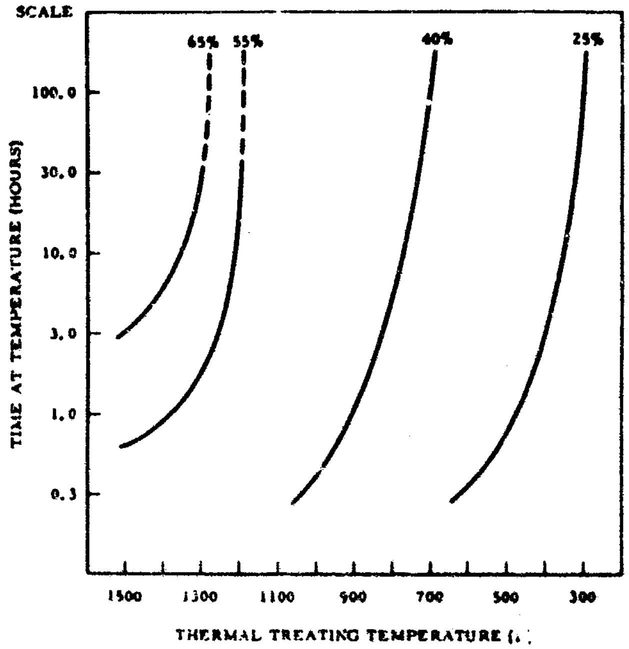
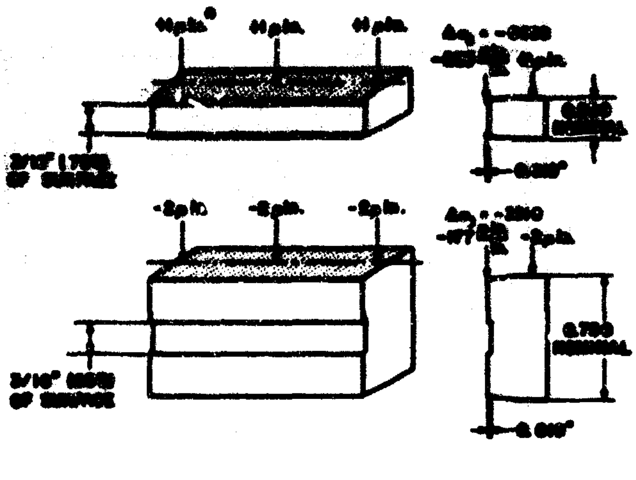


FIGURE 62. EFFECT OF TIME AND TEMPERATURE ON THE RELIEF OF SURFACE STRESS RESIDUAL TO MACHINING IN HOT-PRESSED BERYLLIUM (2% BeO) BASED ON A 95% CONFIDENCE LEVEL EQUAL TO OR GREATER THAN THE NUMBER SHOWN(23)*

*Stress relief can be conducted in air temperatures as high as 900 F as long as appropriate safety precautions are observed. Above 900 F, stress relief should be conducted in an inert atmosphere (such as argon) and in accordance with appropriate safety precautions.



*DEVIATION FROM NOMINAL LENGTH, MICROINCH (μin.)

FIGURE 63. EXPERIMENTAL GAGE BLOCKS AFTER ELECTRO-POLISHING(4)

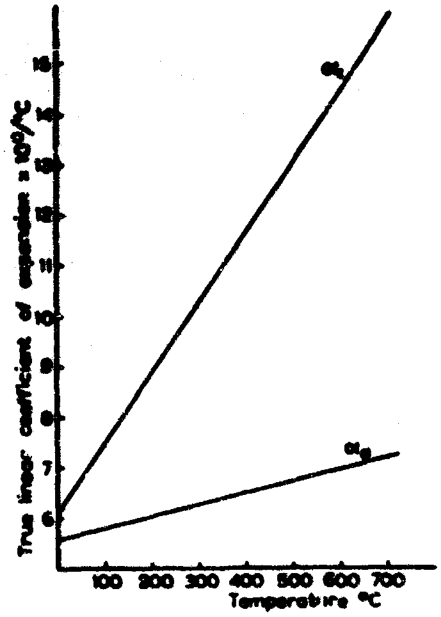


FIGURE 65. EFFECT OF TEMPERATURE ON THE LINEAR COEFFICIENTS OF EXPANSION ALONG THE c AND a AXES OF ALPHA ZIRCONIUM(25)

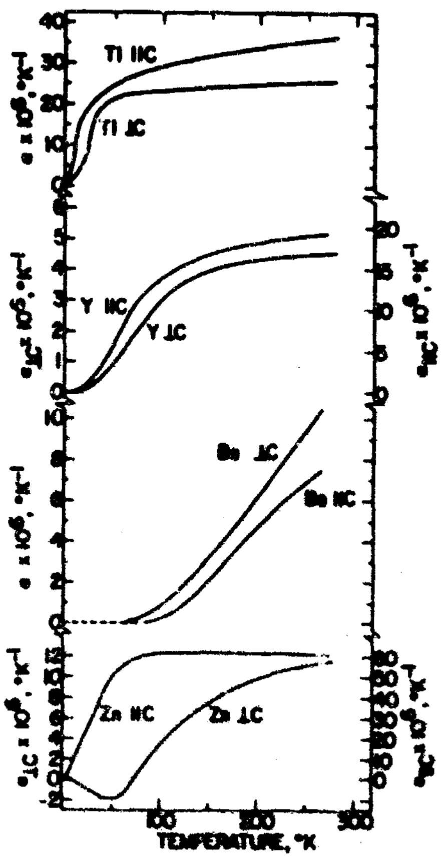


FIGURE 64. TEMPERATURE VARIATION OF THE LINEAR COEFFICIENTS OF EXPANSION(24)

FIGURE 66. MICROSTRUCTURE OF ZIRCONIUM AFTER 5 CYCLES FROM 15 -600 C SHOWING SLIP, BOUNDARY MIGRATION AND A TYPICAL KINK AT A GRAIN-BOUNDARY TRIPLE POINT(25)

Reduced approximately 20 percent in printing

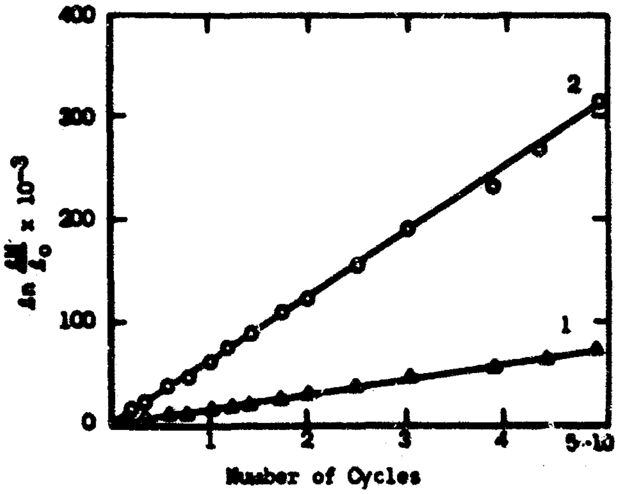


FIGURE 67. DEPENDENCE OF ELONGATION IN ZINC ON THE NUMBER OF THERMAL CYCLES (ΔT = 200)[5]

1 - Forged zinc (55%),
2 - Forged zinc (55%), and then rolled (45%).

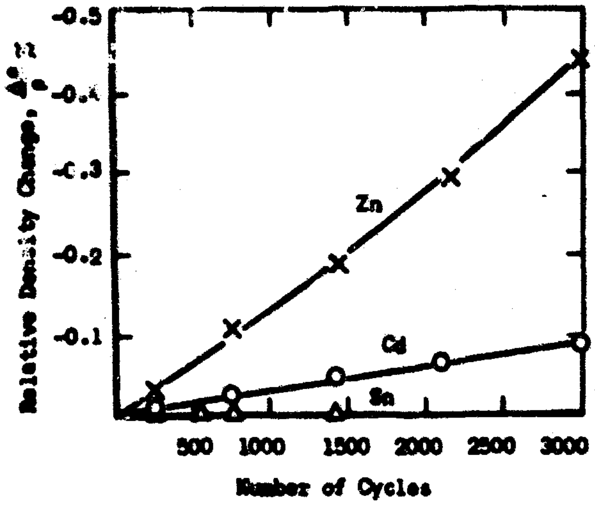


FIGURE 69. DEPENDENCE OF PERCENTAGE CHANGE OF DENSITY ON NUMBER OF THERMAL CYCLES FOR ANISOTROPIC MATERIALS[26]

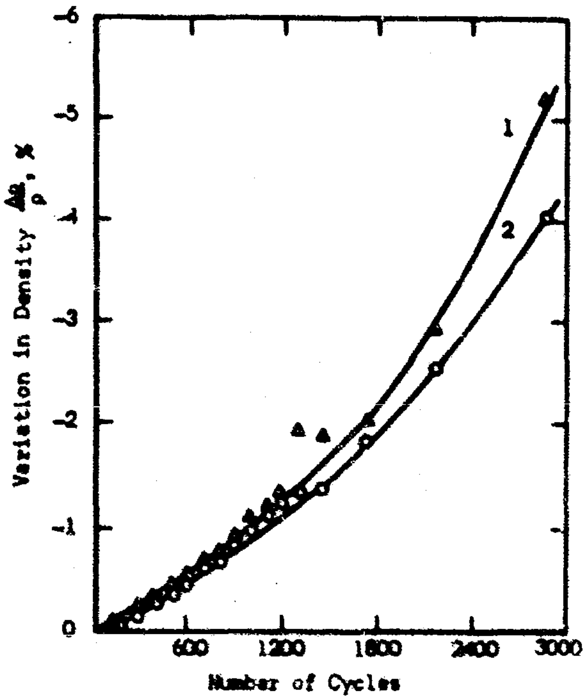


FIGURE 68. DEPENDENCE OF VARIATION IN THE DENSITY OF ZINC ON THE NUMBER OF THERMAL CYCLES (ΔT = 200)[5]

1 - Forged zinc (55%), and then rolled (45%),
2 - Forged zinc (55%).

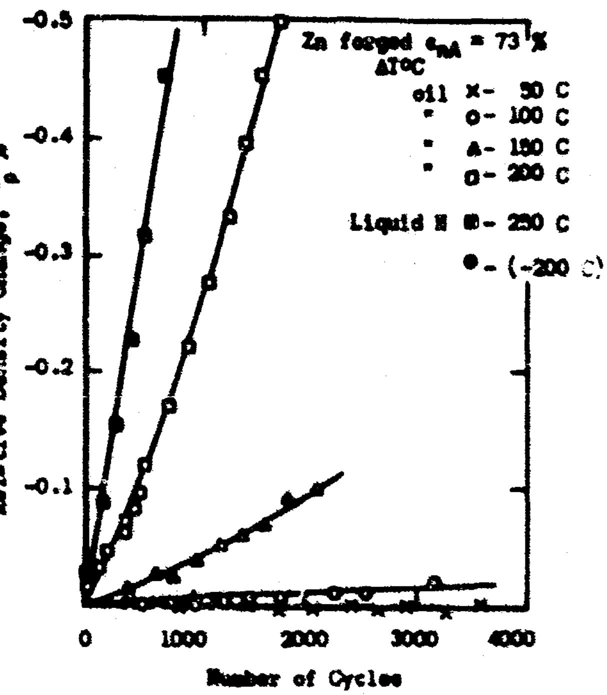


FIGURE 70. PERCENTAGE CHANGE OF DENSITY IN ZINC AT VARIOUS TEMPERATURE RANGES[26]

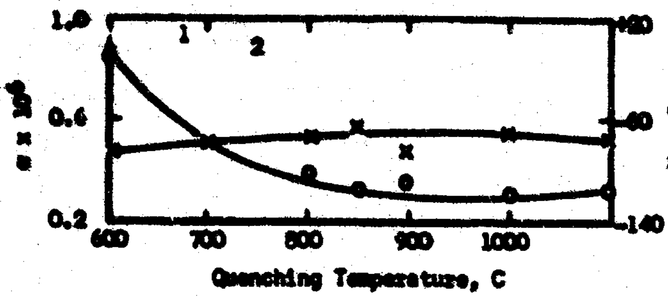


FIGURE 71. VARIATION OF COEFFICIENT OF THERMAL EXPANSION AND TEMPERATURE OF M_H POINT OF E1630A ALLOY (INVAR) WITH QUENCHING TEMPERATURE[27]

1 - Coefficient of thermal expansion,
2 - Temperature of phase transformation.

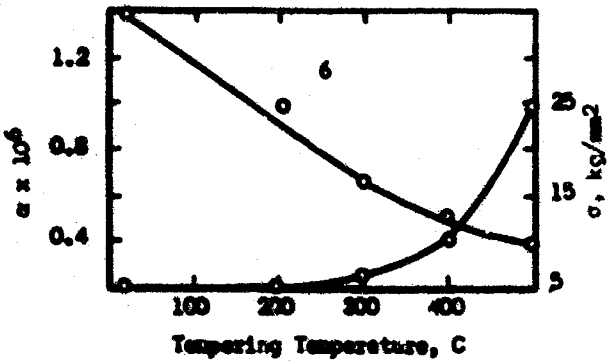


FIGURE 72. VARIATION OF COEFFICIENT OF THERMAL EXPANSION AND RESIDUAL STRESSES (AT SURFACE OF SAMPLE) WITH TEMPERING TEMPERATURE[27]

Quenched from 870 C in water, alloy is Invar.

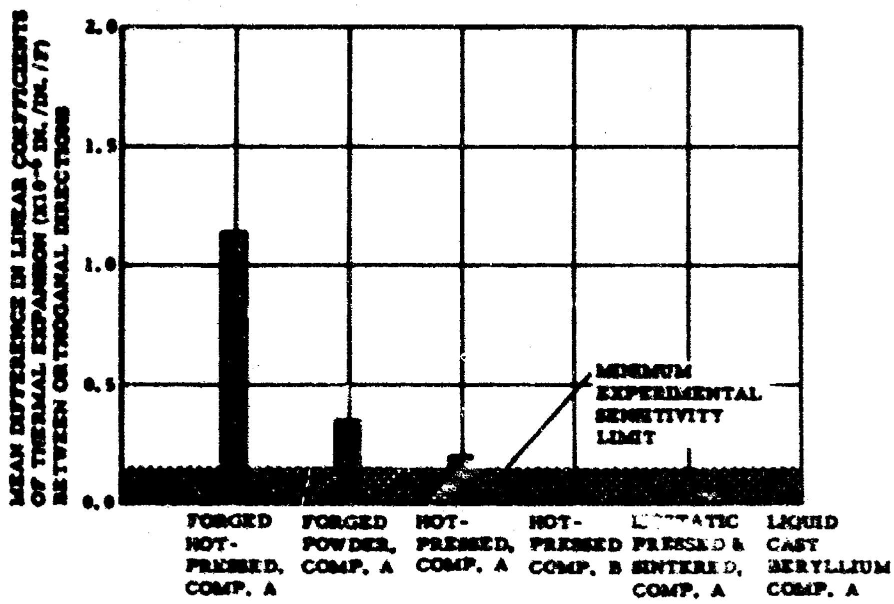


FIGURE 73. COMPARISON OF DEGREE OF THERMAL-EXPANSION ANISOTROPY FOR BERYLLIUM FABRICATED BY SEVERAL TECHNIQUES[28]

(a) Comp A - 2% BeO maximum (b) Comp B - 4% BeO minimum

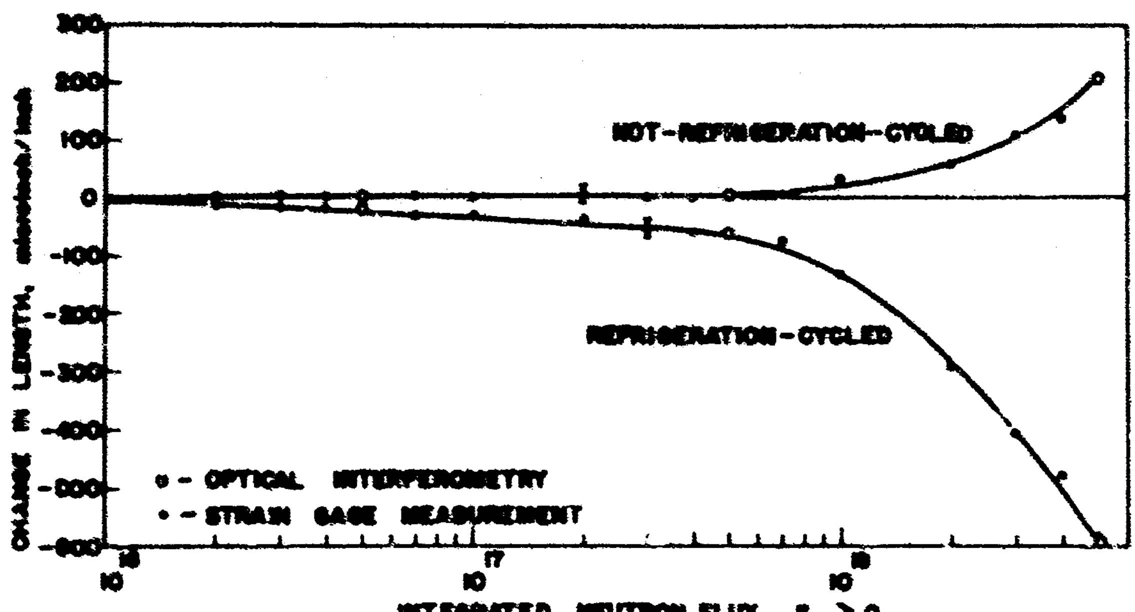


FIGURE 74. CURVES OF INTEGRATED NEUTRON FLUX VERSUS CHANGE IN GAGE BLOCK LENGTH[4]

TABLE 1. PEL COMPARATIVE CHART(18)

Material	Precision Elastic Limit, psi	Ultimate Strength, psi
High-purity alumina	23,000	23,000
High-purity beryllia (ceramic)	20,000	20,000
Glass-bonded mica	4,000	9,000
Hot-pressed beryllium	2,000	45,000
AZ-31B magnesium rod	6,000	34,000
A356-T6 aluminum casting	12,000	40,000
2024-T6 aluminum extrusions	25,000	58,000
6AL-4V titanium rod, an.	70,000	150,000
310 cres steel rod, CR and SR	11,000	120,000
Tool steel, high speed	50,000	180,000
Inconel X	65,000	150,000

TABLE 2. ELASTIC LIMITS AND ELASTIC MODULI OF THREE ALLOYS AT 75, 150, AND 200 F(13)

Material	Test Temperature, F	Elastic Limit, psi	Elastic Modulus, psi
Invar	75	26,400	23×10^6
	150	25,000	22×10^6
	200	24,700	21.5×10^6
356-T6 aluminum	75	8,100	11.3
	150	7,350	10.8
	200	7,190	10.5
310 stainless steel	75	22,700	29.4
	150	20,400	28.5
	200	20,000	28.1
6061 aluminum	75	12,400	10.8
	150	11,700	10.3
	300	10,430	10.0

TABLE 3. MICROMECHANICAL PROPERTIES OF 'A' Ni(a)(10)

Sample	Annealing Temperature, F(b)	σ_o, psi	Grain Size, mm	σ_b, psi	σ_c, psi
1	1200	350	0.018	3500	~20,000
2	1500	<200	0.029	1900	13,000
3	1800	<200	0.081	1800	7,500
4	2300	200	0.280	2000	8,000
5	1800 (aged 100 hr 1200 F)	200	0.081	-	-
6	2300 (aged 100 hr 1200 F)	>200	0.28		
7	2300 (aged 114 hr 1000 F)		0.28	-	-
8	1200	~250	0.048		~10,000
9	1500	210	0.058	3500	7,500
10	1800	200	0.97	2000	7,000
11	2300	200	0.44	1500	6,000
12	2300 (aged 114 hr 1000 F)	200	0.44	1500	~ 5,000
13(c)	2300 (2 hr)	200	0.44	-	-
14(d)	0.5% prestrain	300	0.44	1800	6,000
15	~0.4% prestrain	500	0.44	2500	8,000
16	1.9% prestrain	~300	0.44	3000	7,200
17	4.9% prestrain	~500	0.44	3200	8,000
18	0.3% prestrain	<200	0.44	1400	5,500

(a) See Figure 11 for definition of σ_o, σ_b, and σ_c.
(b) All samples annealed for 1 hr in vacuum unless otherwise indicated. Samples 1-7 rod stock; 8-18 sheet stock.
(c) Sample 13 tested in tension.
(d) Samples 14-18 were annealed 1 hr at 2300 F in vacuum, prestrain, then annealed 1 hr at 1500 F in vacuum prior to testing.

TABLE 4. COMPARISON OF MAIN CONSTRUCTIONAL MATERIALS FOR GYROSCOPES(29)

	Steel (Typical)	Aluminum (Typical)	Beryllium	Ceramic Al_2O_3	Titanium
Density, g/cm^3	7.85	2.7	1.85	3.98	4.51
Young's Modulus, psi	30	10.4	44	56	15
Coefficient of Thermal Expansion, 10^{-6} C	10-12	21-24	11.3	7.0	8.9
Poisson's Ratio	0.28	0.34	0.01	0.205	0.33
Thermal Conductivity, cal $cm^{-2}s^{-1}C^{-1}$	0.10-0.12	0.30-0.57	0.40	0.08	0.04
Ultimate Tensile Strength, psi	60,000-300,000	30,000-90,000	95,000	38,000	80,000-160,000
Precision Elastic Limit, psi	16,000-100,000	12,000-40,000	1,000-12,000	38,000	-
Stiffness/Weight Ratio (steel = 1)	1	1.01	6.25	3.[illegible]	0.87

TABLE 5. STEEL PEL TEST SUMMARY(16)

Alloy Coupon Series	Highlights of Processing History	Average PEL, psi
310 CRES		
S31	Annealed at 2050 F, air blast quench below -900 F, air cool (AC) to room temperature (RT)	5,200
S32	Stress relieve (SR) at 800 F for 24 hrs, furnace cool	10,300
S33	Stress relieve at 800 F for 2 hrs, furnace cool	10,700
440C CRES		
S41	Retarded quench (Q) to RT, SR at 300 F, alternate 3 deep freeze (DF) at -100 F and tempers. Final temper at 650 F.	75,700
S42	Same as S43 except for DF treatments at -320 F	92,200
S43	Same as S41 except DF at -100 F prior to 300 F SR	86,500
52100 air melt		
S51	Marquench, DF at -100 F, alternate 3 tempers at 300 F and DF's	54,600
S52	Same as S51 except for DF treatments at -320 F	61,200
S53	National Bureau of Standards recommended treatment for gage blocks.	40,000
Graph Mo		
S61	Marquench, DF at -100 F, alternate 3 tempers at 300 F and DF's	42,100
S62	Same as S61 except for DF treatments at -320 F	32,800
S63	Same as S61 except final 2 tempers at 650 F and 670 F	60,000

TABLE 6. CHEMICAL COMPOSITION AND MICROSTRUCTURE (%) OF THE INVESTIGATED STEELS(16)

Steel										Composition of Microstructure	
No.	C	Si	Mn	P	S	N	Al	O	Cu	Ferrite	Pearlite
1	0.020	0.12	0.02	0.022	0.027	0.005	0.03	0.012	0.06	100	-
2	0.12	0.13	0.02	0.019	0.007	0.005	0.07	0.010	0.05	91	9
3	0.20	0.18	0.02	0.018	0.008	0.006	0.02	0.015	0.05	86	14
4	0.38	0.20	0.02	0.005	0.014	0.006	0.06	0.011	0.05	67	33
5	0.55	0.18	0.02	0.006	0.013	0.008	0.02	0.011	0.05	49	51
6	0.85	0.17	0.36	0.012	0.010	0.011	0.02	0.006	0.34	-	100
7	0.013	0.13	0.24	0.005	0.008	0.005	0.02	0.020	0.05	100	-
8	0.013	0.11	0.40	0.005	0.008	0.005	0.02	0.028	0.05	100	-
9	0.012	0.08	0.61	0.005	0.009	0.005	0.02	0.023	0.05	100	-
10	0.011	0.10	0.82	0.008	0.007	0.006	0.02	0.018	0.05	100	-

TABLE 7. DIMENSIONAL AND HARDNESS CHANGES RESULTING FROM SUBZERO EXPOSURE OF ANNEALED 17-7PH STAINLESS STEEL[30]

Specimen	After -75 F Exposure	Changes After Indicated Precipitation Heat Treatments: First	Second	Third	Fourth
		Change in Length, mils per inch			
1	+1.06	+0.08	0	0	0
2	+1.02	-0.01	-0.03	-0.03	-0.01
3	+1.23	+0.01	+0.01	0	+0.26
		Change in Diameter, mils per inch			
1	+1.80	0	0	0	0
2	+1.60	+0.16	0	+0.16	0
3	+1.80	0	0	+0.16	+1.60
		Change in Hardness			
1	95 R_B	95 R_B	95 R_B	94 R_B	94 R_B
2	94 R_B	92 R_B	95 R_B	94 R_B	98 R_B
3	95 R_B	97 R_B	96 R_B	20 R_C	26 R_C

Note: (a) Precipitation heat treatment for Specimen 1 -700 F.
(b) Precipitation heat treatment for Specimen 2 -900 F.
(c) Precipitation heat treatment for Specimen 3 -1100 F.
(d) Hardness of all specimens after 1950 F anneal -76 R_B.

TABLE 8. DIMENSIONAL CHANGES IN SAND-CAST Mg-Al-Zn MAGNESIUM ALLOYS OF NOMINAL COMPOSITION[20]

Alloy	Temper	Time, hr	Linear Growth, percent: 212 F	275 F	350 F
AZ63A	-F	24	--	0.001	0.013
		96	--	0.014	0.025
		168	--	0.012	0.025
		720	--	0.025	0.026
	-T4	10	0.001	0.002	0.036
		100	0.003	0.013	0.050
		1,000	0.090	0.050	0.056
		10,000	0.060	0.058	0.058
	-T6	10	0	0	0.002
		100	0.005	0.007	0.015
		1,000	0.007	0.019	0.020
		10,000	0.021	0.020	0.021
AZ92A	-F	24	--	0.008	0.022
		96	--	0.014	0.029
		168	--	0.014	0.031
		720	--	0.034	0.034
	-T4	10	0	0.004	0.047
		100	0.004	0.032	0.070
		1,000	0.028	0.076	0.078
		10,000	0.080	0.078	0.078
	-T6	10	0	0.001	0.008
		100	0.001	0.001	0.024
		1,000	0.004	0.026	0.027
		10,000	0.032	0.027	0.027

TABLE 9. DIMENSIONAL CHANGES IN SAND-CAST AZ91C MAGNESIUM ALLOY OF HIGH AND LOW COMPOSITION[20]

Temper	Time, hr	Linear Growth, percent: 70 F	212 F	250 F	3300 F
		High Composition			
-T4	10	-0.002	-0.001	--	-0.001
	100	-0.003	-0.002	0.017	0.052
	1,000	-0.006	-0.008	0.064	0.077
	5,000	-0.004	0.050	0.082	0.082
	10,000	0.004	0.060	0.082	0.082
	15,000	-0.008	--	--	0.082
	20,000	-0.0005	--	--	0.083
-T6	10	0	0	0	0.003
	100	0	0	0.004	0.020
	1,000	0	0.005	0.028	0.039
	5,000	0	0.018	0.044	0.044
	10,000	0	0.025	0.044	0.046
		Low Composition			
-T4	10	-0.002	-0.002	0.001	-0.011
	100	-0.003	-0.005	0.002	0.012
	1,000	-0.004	-0.007	0.025	0.060
	5,000	-0.004	0.016	0.071	0.072
	10,000	0.004	0.030	0.072	0.074
-T6	10	0	0	0.001	0.003
	100	0	0.002	0.004	0.020
	1,000	0	0.007	0.025	0.057
	5,000	0	0.021	0.068	0.069
	10,000	0.001	0.029	0.068	0.072

TABLE 10. DIMENSIONAL CHANGES IN SAND-CAST MAGNESIUM ALLOYS CONTAINING RARE-EARTH METALS OR THORIUM[20]

Alloy	Temper	Time, hr	Linear Contraction, percent 400 F	500 F	600 F	700 F
EK30A	-T4	10	0.006%	0.003%	0	0
		100	0.007	0.003	0	0
		1,000	0.007	0.003	0	0
	-T6	10	0.006	0.003	0	0
		100	0.007	0.003	0	0
		1,000	0.007	0.003	0	0
EK41A	-T5	10	0.007	0.009	0.009%	0.005%
		100	0.011	0.010	0.010	0.006
		1,000	0.013	0.011	0.010	0.006
		5,000	0.014	0.012	0.011	0.006
	-T6	10	0.012	0.007	0.007	0.002
		100	0.014	0.008	0.008	0.003
		1,000	0.015	0.008	0.008	0.004
		5,000	0.015	0.008	0.008	0.005
EZ33A	-F	10	0.011	0.012	0.015	0.010
		100	0.013	0.015	0.017	0.012
		1,000	0.014	0.018	0.017	0.012
		5,000	0.014	0.019	0.018	0.013
	-T5	10	0.011	0.013	0.012	0.010
		100	0.013	0.016	0.015	0.012
		1,000	0.013	0.018	0.017	0.013
		5,000	0.013	0.019	0.018	0.014
HK31A	-T4	10	0.011	0.010	0.010	0.017
		100	0.011	0.011	0.013	0.018
		1,000	0.011	0.014	—	—
		5,000	0.011	0.014	—	—
	-T6	10	0.003	0.003	0.002	0.006
		100	0.003	0.005	0.005	0.011
		1,000	0.003	0.007	0.013	0.011
		5,000	0.003	0.007	0.013	0.011
HZ32A	-F	10	0.011	0.006	0.006	0.006
		100	0.013	0.007	0.006	0.007
		1,000	0.014	0.008	0.007	0.008
		5,000	0.014	0.008	0.008	0.008
	-T5	10	0.011	0.008	0.008	0.006
		100	0.013	0.010	0.010	0.008
		1,000	0.014	0.011	0.011	0.009
		5,000	0.014	0.012	0.012	0.010

TABLE 11. DIMENSIONAL CHANGES IN SAND-CAST ZK51A-T5 MAGNESIUM ALLOY[20]

Time, hr	Linear Contraction, percent 212 F	300 F	400 F	500 F	600 F
1	0.002%	0.003%	0	0.006%	0.006%
2	0.001	0.005	+0.002	0.005	0.006
4	0.003	0.002	0.001	0.005	0.008
8	0.001	0.006	+0.002	0.005	0.008
24	0.001	0.004	+0.002	0.004	--
48	0	0.004	+0.001	0.005	0.011
96	+0.002	0.002	+0.004	--	--
192	0.001	0.003	0	--	--

TABLE 12. HEAT TREATMENT AND FABRICATION OF 17-4PH STAINLESS STEEL[21]

Procedure[a]	Gage Block Designation P304	P310	P314	P318
Copper plated on nongaging surfaces[b]	-	-	-	X
Austenitized 1925 F	X[c]	X	X	X
Quenched in oil	X[c]	X	X	X
Refrigerated overnight at -140 F	-	X	-	-
Refrigerated 6 hr at -60 F	-	-	X	X
Aged 4 hr at 1025 F	X[c]	-	X	X
Aged 1-1/4 hr at 950 F	-	X	-	-
Refrigerated overnight at -140 F	-	X	-	-
Aged 11 hr at 950 F	-	X	-	-
Ground to size for nitriding	X	X	X	Ends Only
Nitrided at 1020 F, 20% dissociation, approx 45 hr	X	X	X	X
Copper chemically removed	-	-	-	X
Nongaging faces ground, case completely removed	X	X	X	-
Gaging faces ground	X	X	X	X
Stress relieved 4 hr at 1000 F in cracked ammonia	X	X	X	X
Demagnetized	X	X	X	X
Lapped				

(a) Block received only those treatments indicated by X. Order of procedure is chronological from top to bottom.
(b) To inhibit nitriding and eliminate need for grinding of nongaging surfaces after nitriding.
(c) Manufacturer's recommended heat treatment for condition H 1025.

TABLE 13. MIT HEAT TREATMENT FOR BERYLLIUM[31]

General-Use Beryllium

Rough machine all over
Age at 1450 F ±25 in a hydrogen or protective atmosphere for one hour
Furnace cool
Finish machine
Cycle 5 times - from +212 F to -110 F. Air warm and air cool between cycling periods

Instrument-Grade Beryllium

Parts to be rough machined allowing 0.005-in. per side to finish size
Heat to 1450 F ±25 for 1 hour in vacuum or protected atmosphere
Cool at a rate of 1000 per hour maximum to 400 F (vacuum or protected atmosphere not needed below 700 F) then still air cool to room temperature

Stabilize

Machine parts to finish size
Heat to +210 F ±10 in deionized boiling water for 10 minutes
Remove to room temperature for 10 minutes
Cool to -100 F ±10 in acetone and dry ice (for 10 minutes)
Repeat procedure 5 cycles.

TABLE 14. EFFECT OF TEMPERATURE CHANGES ON THE STRESS BUILD-UP IN POLYCRYSTALLINE, NONCUBIC MATERIALS(5)

Material	Lattice Type	α_0, $\frac{kg}{mm^2\ degree}$*	ΔT, C	$\alpha_0, \Delta T\ \frac{kg}{mm^2}$	$\frac{psi}{C}$
Uranium	Monoclinic (orthorhombic)	0.254	550	135.0	360
Selenium	Hexagonal	0.154	200	30.8	219
Zinc	Hexagonal	0.125	400	50.0	178
Cadmium	Hexagonal	0.0645	300	19.3	92
Tin	Tetragonal	0.0506	200	10.1	72
Tellurium	Hexagonal	0.0322	350	11.3	46
Zirconium	Hexagonal	0.0272	850	24.0	39
Antimony	Rhombohedral (trigonal)	0.0150	600	9.0	21
Bismuth	Rhombohedral (trigonal)	0.0066	250	1.65	9.4
Magnesium	Hexagonal	0.00193	550	1.06	2.7

* α_0 is the maximum stress which may arise close to the grain boundaries in a polycrystalline aggregate with variation of temperature by 1°; ΔT the possible ranges of temperature; $\alpha_0\ \Delta T$ the stresses corresponding to this range of temperatures.

TABLE 15. EFFECT OF CRYOGENIC AGING AND REACTOR IRRADIATION ON THE LENGTH OF GAGE BLOCKS AS MEASURED BY OPTICAL INTERFEROMETRY(4)

Specimen Designation	Treatment	Change in 0.750 Inch Length, ±2.0 microinches/in.
	No Refrigeration Cycled	
Control	Room-temperature storage	+1.0
Lot 100	Cryogenically aged	+32.0
Lot 101	Irradiated 5×10^{16} nvt	+2.0*
Lot 102	Irradiated 5×10^{17} nvt	+3.0*
Lot 103	Irradiated 5×10^{18} nvt	+207.0*
	Refrigeration Cycled	
Control	Room-temperature storage	+2.0
Lot 120	Cryogenically aged	-3.0
Lot 121	Irradiated 5×10^{16} nvt	-15.0*
Lot 122	Irradiated 5×10^{17} nvt	-53.0*
Lot 123	Irradiated 5×10^{18} nvt	-591.0*

* Measurement taken 7 days after end of irradiation.

TABLE 16. EFFECT OF CRYOGENIC AGING AND REACTOR IRRADIATION ON THE AMOUNT OF RETAINED AUSTENITE IN GAGE BLOCKS(4)

Specimen Designation	Treatment	Amount of Retained Austenite, ±0.3 vol% Before, A_i	Change, ΔA	$\frac{\Delta A}{A_i}$ x 100%
	Not Refrigeration Cycled			
Control	Room-temperature storage	8.9	-0.2	-2.2
Lot 100	Cryogenically aged	10.6	-0.9	-8.5
Lot 101	Irradiated 5×10^{16} nvt	13.5	[illegible]	-1.5
Lot 102	Irradiated 5×10^{17} nvt	9.9	-0.3	-3.1
Lot 103	Irradiated 5×10^{18} nvt	14.5	-1.6	-11.0
	Refrigeration Cycled			
Control	Room-temperature storage	3.6	-0.1	[illegible]
Lot 120	Cryogenically aged	4.6	-0.6	-13.0
Lot 121	Irradiated 5×10^{16} nvt	5.3	+0.1	+2.0
Lot 122	Irradiated 5×10^{17} nvt	3.2	-0.6	[illegible]
Lot 123	Irradiated 5×10^{18} nvt	4.7	[illegible]	-38.4

TABLE 17. EFFECT OF CRYOGENIC AGING AND REACTOR IRRADIATION ON THE SUPERFICIAL-RESIDUAL MACROSTRESS IN GAGE BLOCKS(4)

Specimen Designation	Surface	Treatment	Residual Macrostress* Before ±4000 psi	Residual Macrostress* Change ±8000 psi
		Not Refrigeration Cycled		
Control	Side 1	Room-temperature	-48,700	-3,700
	Side 2	storage	-54,200	-2,500
Lot 100	Side 1	Cryogenically	-68,400	-800
	Side 2	aged	-71,200	-16,450
Lot 101	Side 1	Irradiated	-34,200	-2,900
	Side 2	$5x10^{16}$ nvt	-38,600	-3,700
Lot 102	Side 1	Irradiated	-56,350	-2,300
	Side 2	$5x10^{17}$ nvt	-64,700	-1,900
Lot 103	Side 1	Irradiated	-69,000	-2,200
	Side 2	$5x10^{18}$ nvt	-58,700	-2,600
		Refrigeration Cycled		
Control	Side 1	Room-temperature	-114,700	-2,100
	Side 2	storage	-105,200	-3,600
Lot 120	Side 1	Cryogenically	-133,800	-24,700
	Side 2	aged	-140,200	-4,400
Lot 121	Side 1	Irradiated	-163,400	-8,800
	Side 2	$5x10^{16}$ nvt	-156,300	-10,300
Lot 122	Side 1	Irradiated	-144,300	-21,600
	Side 2	$5x10^{17}$ nvt	-138,200	-17,300
Lot 123	Side 1	Irradiated	-123,600	-33,250
	Side 2	$5x10^{18}$ nvt	-133,200	-29,600

* Minus sign (-) indicates compressive stress as measured on martensite (112)-(211) unresolved doublet.

TABLE 18. EFFECT OF ROOM-TEMPERATURE AGING ON THE LENGTH OF IRRADIATED-GAGE BLOCKS(4)

Specimen Designation	Irradiation Exposure, En <0	Deviation from Nominal 0.750 In. Length, ±2.0 microinches/in.			
		Preirradiation	Time After Irradiation 7 days	14 days	21 days
		Not Refrigeration Cycled			
Control	None	+1.0	+2.0	+2.0	+2.0
Lot 101	$5x10^{16}$ nvt	-1.0	+9.0	+7.0	+9.0
Lot 102	$5x10^{17}$ nvt	-2.0	+3.0	+5.0	+10.0
Lot 103	$5x10^{18}$ nvt	+1.0	+207.0	+256.0	+258.0
		Refrigeration Cycled			
Control	None	0.0	+1.0	+1.0	+1.0
Lot 121	$5x10^{16}$ nvt	0.0	-15.0	-12.0	-11.0
Lot 122	$5x10^{17}$ nvt	-1.0	-53.0	-45.0	-49.0
Lot 123	$5x10^{18}$ nvt	0.0	-591.0	-554.0	-579.0

TABLE 19. SUMMARY OF CHANGES OF LENGTH, SUPERFICIAL RESIDUAL MACROSTRESS, AND AMOUNT OF RETAINED AUSTENITE IN GAGE BLOCKS(4)

Specimen Designation	Treatment	Change in Length, ±2.0 μin./in.	Averaged Change in Residual Stress, ±8000 psi	Change in Amount of Retained Austenite, ±0.3 vol%
		Not Refrigeration Cycled		
Control	Room-temperature storage	+1.0	-3,100	-0.2
Lot 100	Cryogenically aged	+32.0	-8,600	-0.9
Lot 101	Irradiated $5x10^{16}$ nvt	+2.0	-3,300	-0.2
Lot 102	Irradiated $5x10^{17}$ nvt	+3.0	-2,200	-0.5
Lot 103	Irradiated $5x10^{18}$ nvt	+207.0	-2,400	-1.6
		Refrigeration Cycled		
Control	Room-temperature storage	+2.0	-2,850	-0.3
Lot 120	Cryogenically aged	-3.0	-14,550	-0.6
Lot 121	Irradiated $5x10^{16}$ nvt	-15.0	- 9,500	+0.1
Lot 122	Irradiated $5x10^{17}$ nvt	-53.0	-19,450	-0.6
Lot 123	Irradiated $5x10^{18}$ nvt	-591.0	-31,000	-1.8

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13. ABSTRACT

This memorandum discusses some of the problems that arise as a result of dimensional instability, and presents data on stability, precision mechanical properties, and stabilization procedures for a variety of materials. The memorandum is intended to supplement DMIC Memorandum 169, "A Review of Dimensional Instability in Metals", by F. C. Holden, and dated March 19, 1964. Emphasis is placed on the causes and effects of dimensional instability. These are discussed, and available information on how they can be controlled is presented.

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Dimensional instability	8	3				
Plastic deformation	8	3				
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Metallurgical mechanisms	8	3				
Physical properties	8	3				
Stress strain	7,8	3				
Precision elastic limit	8	3				
Microyield stress	8	3				
Stress-microstrain	8	3				
Creep	8	3				
Microcreep	8	3				
Temperature	6	3				
Coefficient of expansion	7	3				
Zirconium	1	3				
Nickel	1	3				
Iron-base alloys	1	3				
Copper	1	3				
Aluminum alloys	1	3				
Beryllium	1	3				
Stainless Steel	1	3				
High-strength steel	1	3				
Superalloys	1	3				
Magnesium alloys	1	3				
Titanium	1	3				

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